NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

TECHNICAL NOTE

No. 1465

THE RUPTURE-TEST CHARACTERISTICS OF HEAT-RESISTANT SHEET ALLOYS AT 1700° AND 1800° F

By J. W. Freeman, E. E. Reynolds, and A. E. White

University of Michigan

Washington
February 1948

NATIONAL ADVISORY COMMITTEE FOR AERONAUTICS

TECHNICAL NOTE NO. 1465

THE RUPTURE-TEST CHARACTERISTICS OF HEAT-RESISTANT SHEET ALLOYS AT 1700° AND 1800° F

By J. W. Freeman, E. E. Reynolds, and A. E. White

SUMMARY

The characteristics of a representative group of heat-resistant alloys in sheet form were investigated by rupture tests at 1700° and 1800° F. The primary purpose was to provide data which could be used by the aircraft industry to determine whether the rupture-test properties of an alloy would be a criterion of service performance in combustion chambers and other applications of sheet alloys at high temperatures in power plants for aircraft. Suitable service data were not available for this type of correlation in the present report.

The materials studied included the standard chromium-nickel types 330 (35Ni-15Cr), 310 (25Cr-20Ni), 310S (25Cr-20Ni+Si), 309 (25Cr-12Ni), and Inconel alloys, the new highly alloyed materials known as Vitallium, Co-Cr-Ni, S816, S590, and low-carbon N-155, and four experimental alloys containing cobalt, molybdenum, tungsten, and boron in addition to nickel and chromium. The highly alloyed materials, in most cases, had higher rupture strengths only for time periods up to about 1000 hours at 1700° F and for somewhat shorter time periods at 1800° F. The relative rupture strengths varied with the time period and the temperature considered. In general, Vitallium, J-838, and S-590 were the strongest of the new alloys and type 310S was the best of the standard alloys. Low-carbon N-155 apparently had the best strength at time periods of 1000 hours and longer.

The rupture strengths at 1700° F ranged up to 12,000, 8000, 5200, and 3900 psi for fracture in 10, 100, 400, and 1000 hours, respectively. The corresponding range in rupture strengths at 1800° F was up to 8600, 4200, 2500, and 2000 psi.

Rather wide differences in properties of separate heats of type 310S and Inconel alloys showed that treatment prior to testing influenced the rupture-test characteristics. The apparent equalization of rupture strengths at the longer time periods was attributed to agglomeration, solution, or lack of precipitation of the excess constituents believed to impart high rupture strength to some of the alloys at lower temperatures. Apparently the strengths of the various matrix materials were not greatly different. Service conditions under which corrosion would differ from that during the rupture tests would result in different rupture properties. Presumably a less-oxidizing condition would improve the performance by some of the alloys.

INTRODUCTION

One of the difficulties encountered in the operation of gas turbines has been the short life of liners for combustion chambers. A suitable metallurgical test which would give results correlating with service performance was not available. It was suggested that the rupture test might fulfill the need.

Accordingly, an investigation of the characteristics of a representative group of alloys was conducted to provide rupture strength and ductility data. Fourteen alloys in the form of 0.031-inch-thick sheet were used. A number of these were obtained from commercial production and the others were experimental alloys. The range of service temperatures was estimated at 1700° to 1800° F and the tests were confined to these temperatures.

Suitable data were not available during preparation of this report to correlate the performance of the alloys in service with the rupture-test characteristics. In an evaluation of the test and the alloys, due consideration should be given to the possibility that the data may not be typical for some of the alloys. There is reason to believe that the rupture-test characteristics vary to a considerable extent, depending on the prior treatment during manufacturing for any one alloy. The influence of this factor has not been determined for sheet alloys.

The investigation was conducted under the sponsorship and with the financial assistance of the National Advisory Committee for Aeronautics as a part of their program of sponsored research at the Department of Engineering Research of the University of Michigan on heat-resisting alloys for gas turbines. The program has covered extensive work on rotor materials and is being continued in this field and in the field of sheet alloys for combustion chambers and other high-temperature-duct work.

TEST MATERIALS

Fourteen alloys, the chemical compositions of which are shown in table I, were considered in the investigation. Five of these, types 330, 310, 310S, 309, and Inconel are well known chromium-nickel-iron alloys which have been used widely for heat-resisting applications for a number of years. Five of the alloys, Vitallium, Co-Cr-Ni, S816, S590, and low-carbon N-155 are relatively new alloys containing large amounts of cobalt, chromium, and nickel, along with additions of tungsten, molybdenum, or columbium. The four alloys, J-837, J-838, J-839, and J-840, were new experimental alloys.

Two separate lots of Inconel and type 310S alloys were included. This was done primarily to obtain some measure of the degree to which the data reported are typical. The particular alloys were selected because they were of special interest for combustion liners and other high-temperature-duct work in gas turbines.

All the test materials were supplied in a condition in which they ordinarily would be furnished commercially for high-temperature service. Possible exceptions are that this might be altered by further experience with the new highly alloyed materials and the four experimental heats. In fact, six of the alloys were obtained from a consumer who had purchased the materials for the construction of combustion chambers.

The materials were supplied as strips 22 inches long, $1\frac{1}{4}$ inches wide, and 0.030 to 0.040 inch thick. The width in the 2-inch gage test section was reduced to 1 inch and the specimens were tested as such. The available details of processing and heat treatment are summarized as follows:

Vitallium:

The material tested was from Crucible melts. Ingots were forged between 2040° F and 1900° F. Rolling of the hammered slabs was done from 2100° to 1900° F. After finish rolling, the pieces were annealed 6 minutes at 2050° F and pickled. This material was furnished by the General Electric Company.

Co-Cr-Ni:

The sheets were hot-rolled with a finishing temperature of 2100° to 2125° F and then annealed by heating at 2175° to 2200° F. In the annealing process, the material was held in the furnace just long enough to reach a required temperature and then air-cooled. The Haynes Stellite Company made the sheets, and the specimens were furnished by the General Electric Company.

S816:

Sheets were hot-rolled to 0.035-inch gage at 2200° to 2250° F, annealed at 2200° to 2250° F for 5 minutes and air-cooled. The material was then sand-blasted, scrubbed, given two passes on cold rolls, and buckled. It was made by the Allegheny-Ludlum Steel Corporation and furnished by the General Electric Company.

S590:

The final processing included annealing at 2150° F with an air cool. The sheets were then sand-blasted, acid-dipped, scrubbed, and buckled. The samples were made and supplied by the Allegheny-Ludlum Steel Corporation.

Inconel:

The materials designated as Inconel-1 and Inconel-2 were hot-rolled with a final rolling temperature of 2100° F and annealed at 1900° F, followed by a pickle. Sheets were processed by the International Nickel Company. Tests on Inconel-2 were made from a single sheet.

Low-carbon N-155, J-837, J-838, J-839, and J-840 alloys:

Ingots from a 30-pound induction furnace heat were hot-forged to sheet bars approximately 2 inches wide by 1 inch thick, and the bars were rolled to strips. Forging was from approximately 2100° to nearly 1500° F. The sheets were rolled from approximately 2000° to nearly 1400° F. Samples were then annealed at 2000° to 2100° F and air-cooled. These alloys were made and supplied by the Union Carbide and Carbon Research Laboratories, Inc.

Type 330:

Sheets were hot-rolled from a large induction heat, annealed at 1950° F, air-cooled, and pickled. The Crucible Steel Company made and supplied the specimens.

Type 310:

The Crucible Steel Company fabricated and supplied specimens from a large induction heat. The sheets were hot-rolled, annealed at 2050° F, air-cooled, and pickled.

Type 310S:

The material designated AF-18 was hot-rolled to 0.045-inch gage sheet from 2100° F, annealed for 6 to 8 minutes at 2100° to 2150° F and air-cooled, sand-blasted, spot-ground, cold-rolled to 0.033-inch gage, annealed 2100° to 2150° F for 6 to 8 minutes and air-cooled, sand-blasted, scrubbed, cold-rolled one pass, and buckled. This alloy was made by the Allegheny-Ludlum Steel Corporation and supplied by the General Electric Company.

The material from heat 14626 was hot-rolled to 0.050-inch gage sheet, pickled, annealed at 2180° F for 6 minutes, water-quenched, sand-blasted, scrubbed, cold-rolled to 0.040 inch, annealed 2180° F for 6 minutes, water-quenched, sand-blasted, scrubbed, cold-rolled one pass to flatten, roller-leveled, and sheared. This alloy was made and supplied by the Allegheny-Ludlum Steel Corporation.

Type 309:

No processing information is available for this material. Presumably it was hot-rolled and annealed. It was furnished by the General Electric Company.

PROCEDURE

This investigation of sheet materials included sufficient tests to evaluate the rupture properties at 1700° and 1800° F from 1 hour to at least 400 hours. The alloys were examined metallographically before and after rupture testing.

The short-time tensile tests were run on those alloys for which a sufficient number of specimens were supplied. The tensile machine was operated at a head speed of 0.03 inch per minute after holding the specimen at the test temperature for 1 hour. No strain data were obtained during these tests. Rupture tests of less than 10-hour duration were also conducted in the tensile machine after holding the specimen for 1 hour at temperature.

The longer-time rupture tests were run in single-specimen units. The specimens were held approximately 24 hours at the required temperature prior to application of the load. Direct loading was used to apply the stress to the lower-stress tests. The higher stressed specimens were loaded by a simple beam through a system of knife edges. No time-elongation data were obtained from the rupture tests because of the difficulty of making reliable measurements on thin sheets, particularly at these temperatures.

Metallographic samples of the cross section, taken lengthwise of the strips and including the fracture of the rupture specimens from the longest duration tests at each temperature, were polished, etched, and photographed at magnifications of 100X and 1000X.

RESULTS

The data from the rupture and tensile tests are given in table II. The usual logarithmic curves of stress against rupture time are shown in figure 1. The relative rupture strengths (obtained from fig. 1) and ductility varied with both the time period and the temperature, as is shown by the tabulation of rupture strengths and estimated ductility in tables III and IV and their graphical comparison in figure 2.

A tendency toward erratic results was characteristic of some of the materials. In some cases these stress-rupture time curves (fig. 1) showed abrupt changes in slope. The curves for type 330 at 1700° F and J-840 at 1800° F showed an unusual characteristic in that they had a lower slope at longer time periods.

The stress for rupture in 100 hours varied from 1950 to 8000 psi at 1700° F and 2200 to 4200 psi at 1800° F. In general, Vitallium, J-838, S590, and low-carbon N-155 had the highest rupture strengths. Type 330 was the weakest at 1700° F; whereas, Inconel and type 309 were only slightly stronger. The weaker alloys at 1700° F were not tested at 1800° F. Of those tested at 1800° F, S816 had the lowest strengths for time periods longer than 100 hours.

The degree of ductility to fracture in the rupture tests varied to a considerable extent between alloys. Of the alloys with the highest rupture strengths, J-838 had the highest elongation, although the others also had good ductility. Low elongation was exhibited by Co-Cr-Ni and type 310S alloys.

The low-nickel alloys J-837 and J-839 were subject to rapid oxidation and failed from this cause in the rupture tests. The loss in section size due to oxidation was of such influence that it was not possible to obtain the usual stress-rupture time curves. The low oxidation resistance for these two alloys was verified by heating small samples without stress at 1700° F and obtaining complete oxidation in 100 hours.

The original materials and the fractured rupture specimens from the most prolonged tests were examined metallographically. The information obtained is illustrated by the photomicrographs of figures 3 to 15 and is summarized as follows:

(1) There was no appreciable tendency for grain growth in any of the alloys during rupture testing. There was, however, considerable variation in grain size between individual alloys.

(2) A layer of clear structure was observed at the surface of the rupture specimens of all the alloys. This was probably the result of decarburization or a loss of other alloying elements by a similar mechanism.

(3) Agglomeration of the excess constituent occurred during rupture testing in most of the materials. This was accompanied by further precipitation in Co-Cr-Ni and type 310S alloys. These were also the only two alloys in which intergranular precipitation occurred.

(4) There was a considerable degree of difference in the amount of intergranular cracking which occurred near the fracture of the various materials and in the amount of oxidation that occurred in these cracks. Inconel, in particular, was very susceptible to intergranular oxidation.

DISCUSSION OF RESULTS

The primary purpose of this investigation was to provide rupture-test data for a representative group of alloys which might be used for combustion-chamber liners and other high-temperature applications for sheet materials in power plants for aircraft. Data at 1700° and 1800° F are presented for several of the standard chromium-nickel alloys, some of the newer, highly alloyed materials, and some new experimental modifications.

The newer alloys were, in general, outstandingly stronger than the standard chormium-nickel alloys only at 1700° F. This degree of superiority decreased as the time for rupture increased. The same general trends existed at 1800° F, where the time period for equalizing rupture strengths was less than at 1700° F. One heat of the standard chromium-nickel alloy, type 310S with 0.13 percent carbon, had considerably higher rupture strengths than the other standard alloys. This particular heat compared quite favorably in strength with the new, highly alloyed materials. One heat of Inconel alloy had an exceptionally high 1000-hour strength at 1800° F. One of the more highly alloyed materials, S816, seemed to have abnormally low strength.

Other factors than chemical composition certainly influenced the variations in strength between the alloys. The differences in rupture strength between the two heats of type 310S and between the two lots of Inconel suggest that processing procedure and heat treatment had a pronounced influence on the results. Oxidation probably influenced the results as was indicated by the tests of Inconel, J-837, and J-839 alloys.

The possible effects of processing procedure and heat treatment may be appraised by the comparison of data in table V for the alloys in sheet, bar stock, and cast forms. A rather wide range in strength has been obtained for each alloy. In general, the cast form had the highest strength, the bar stock intermediate, and the sheet form lowest. There were, however, exceptions, depending on the heat treatment of the bar stock. For instance, low-carbon N-155 in the hot-rolled condition and types 309 and 310 with lower-temperature heat treatments were weaker than the sheet materials.

Oxidation undoubtedly influenced the results. This influence may have taken place in two ways. One is that some alloys, such as type 310S, had better strength because of high resistance to oxidation. At the opposite extreme were J-837 and J-839, which failed primarily by oxidation. The other possible effect would be that part of the low strength of sheet materials in comparison with bar stock was due to a larger surface area per unit of cross section exposed to oxidation than in the round specimens from bar stock.

A recent study of the results of rupture tests of standard chromium-nickel alloys at temperatures up to 1800° F suggested that nitrogen contamination, as well as oxidation, might have influenced the results of this investigation. (See reference 3.) In reference 3, nitrogen contents as high as 1.33 percent were observed after rupture tests at 1800° F on 25Cr-20Ni steel. The microstructures of the type 310 and type 310S sheet samples suggested the possibility of a nitrogen pickup in this investigation. Accordingly several samples were analyzed for nitrogen with the following results:

Alloy	Test conditions			Nitrogen content near fracture (percent)
	Temperature ($^{\circ}F$)	Stress (psi)	Rupture time (hr)	
S816	1800	1500	260	0.27
Type 330	1700	1500	811	.23
Type 310	1700	2500	728	.19
Type 310S (AF-18)	1700	3000	1124	.26
	1800	2200	384	.22
Type 310S (14626)	1700	2400	397	.12

While the nitrogen contents of the original test materials were not available for comparison, certainly these results do not indicate the pronounced nitrogen pickup shown by reference 3. Sampling may have influenced these results, however, because reference 3 reported the nitrogen content to be highest at fracture of the specimens. There may have been undue dilution of the metal adjacent to the fractures in the sheet samples in order to obtain a sufficiently large sample for the analysis. Nitrogen pickup may therefore have been a larger influence on the results of the rupture tests in this investigation than is indicated by the foregoing analyses.

The primary reason why the newer alloys lost their superiority at longer time periods and in less time at 1800° F than at 1700° F was probably structural instability. These alloys are believed to derive their high strength at lower temperatures from precipitation of complex compounds. The microstructures of the samples indicate that these complex compounds agglomerate, redissolve, or do not precipitate at 1700° and 1800° F. Consequently, the strengthening effect from these particles, which is normal at lower temperatures, is lost or does not occur. The decarburization, or other alloy loss from the surface of the specimens, is another structural instability that may, in general, have adversely affected strength in the rupture test.

In summarization, then, it appears that the properties obtained from the rupture tests are dependent on strength of the matrix material, prior heat treatment, structural stability, and surface stability of the alloys under the test conditions. It follows that the properties might be considerably different under conditions in which the surface attack would be different from the air atmosphere of the tests. It might well be that a combustion-gas atmosphere with reducing characteristics would considerably increase the load-carrying ability.

The behavior of some of the alloys in the tests are specific examples of the general cases previously discussed. For example:

(1) Inconel appeared to be particularly susceptible to intergranular oxidation under the test conditions. (See fig. 7.) This oxidation was apparently more severe during testing at 1700° than at 1800° F. The erratic rupture time periods at 1700° F and the indication of equal strengths at 1700° and 1800° F for long time periods were probably due to oxidation rather than alloy strength controlling the time for rupture. The similarity of microstructure of both heats also indicates that oxidation rather than alloy strength controlled the rupture times.

(2) The low ductility of Co-Cr-Ni and type 310S alloys was assiciated with structural instability of the intergranular precipitation type. (See figs. 4, 13, and 14.) Apparently the nearly continuous envelopes of excess constituents which developed in the grain boundaries caused some brittleness. These were also the only alloys in which an appreciable amount of precipitation occurred during the tests.

(3) The structures of the type 310S samples, especially heat 14626 at 1700° F, were unusual in that the shape and the distribution of excess constituents were quite different from the other materials. (See fig. 14.) The amount of these constituents also seems greater than would be anticipated on the basis of the reported analyses.

(4) In type 310 alloy an easily etched constituent formed in the grain boundaries, but not in a continuous network. (See fig. 12.) Either this constituent does not cause brittleness or its discontinuous nature does not interfere with ductility at 1700° F.

(5) The structures of the more highly alloyed materials S816, S590, low-carbon N-155, J-838, and J-840, were quite similar. (See figs. 5, 6, 8, 9, and 10.) Testing caused relatively little change in the structures, although varying amounts of oxidation and intergranular cracking occurred. There was some indication that S816 underwent more structural alteration and oxidation than the other alloys of this group. The structure of the Vitallium sheet was comparable with these alloys, although both the grain size and the size of the excess constituent particles were larger. (See fig. 3.)

(6) All the alloys were subject to intergranular cracking during the rupture tests. No consistent relation, however, between this phenomenon and properties of the alloys could be observed.

The solution effects for the various annealing treatments varied to a considerable extent. Complete solution resulted in type 310S (heat 14626) and type 309. Either the temperature was too low, or insufficient time was allowed to dissolve carbides in Co-Cr-Ni, type 310, the AF-18 specimens of type 310S, and type 330 alloys. The newer materials with the higher alloy contents contained large particles of excess constituents which made it difficult to judge the relative solution effects. These particles apparently acted to maintain a fine grain size during annealing.

Most experience with the behavior of materials at high temperatures has indicated that at the higher temperatures a larger grain size gives higher strength than a fine grain size. The two heats of type 310S had different grain sizes. The finer-grained material, AF-18, however, had the higher strength. The usual effect of grain size may have been masked by the previously described complex structural changes in these alloys.

There are several factors which influence the properties of the alloys considered in this investigation. For this reason, there is a reasonable doubt whether the variations in properties which seemingly are due to chemical composition are due to this cause alone. For instance:

(1) The difference in structural characteristics between the two heats of type 310S suggests that heat treatment, instead of the variation in carbon content, may have been responsible for the difference in strengths.

(2) Apparently the higher cobalt of S816 in comparison with S590 was responsible for lower strength. The comparative data for bar stock and castings in table V indicate, however, that S590 is not universally better and that some other factor is also influencing the results. The description of the fabrication of these two alloys suggests that the S816 sheets were cold-worked after the final anneal, which would increase structural instability under the testing conditions. The relative effectiveness of heat treatments or differences in chromium content may also have been factors that influenced the rupture properties.

(3) In alloys J-837, J-838, J-839, and J-840 it appears that the lower nickel content of J-837 and J-839 was responsible for low oxidation resistance. The indication that the higher nitrogen content of J-840 in comparison with J-838 was detrimental to high-temperature properties should be checked on other heats before it is accepted as the controlling factor.

(4) In some cases there was remarkably little difference in rupture properties of alloys with rather specific variations in composition. For instance J-838 and J-840 differed from S590 in that they did not contain the 4 percent of columbium of S590 and their nickel content was 5 percent less. Boron was also added to J-838 and J-840, yet their rupture properties were similar.

CONCLUSIONS

Rupture tests conducted at 1700° and 1800° F on a representative group of alloys in sheet form led to the following conclusions:

1. The degree of superiority of the highly alloyed new materials over the standard chromium-nickel-iron alloys decreases with time for rupture at 1700° F, so that at time periods in excess of 1000 hours there are only relatively small differences. At 1800° F the time period for near equalization of strengths is considerably less than at 1700° F.

2. Sheet Vitallium, S590, low-carbon N-155, and J-838 alloys gave the highest strengths for most of the time periods considered. Type 310S was the best of the standard alloys.

3. Separate lots of type 310S and Inconel alloys in sheet form had quite widely varying properties, suggesting that prior heat treatment is of considerable influence on the properties. This was substantiated by data from other investigations.

4. Oxidation during the tests influenced the rupture test characteristics, indicating that the properties might depend considerably on the corrosion conditions during service. Less oxidizing conditions would probably increase rupture strength.

5. The tendency toward equalization of rupture strengths, regardless of alloy content, appears to be due to agglomeration, solution, or lack of precipitation of the excess constituent believed to control strength at lower temperatures.

Department of Engineering Research
University of Michigan
Ann Arbor, Mich., August 26, 1946

REFERENCES

1. Freeman, J. W., Reynolds, E. E., and White, A. E.: The Rupture Test Characteristics of Six Precision-Cast and Three Wrought Alloys at 1700° and 1800° F. NACA ARR No. 5J16, 1945.

2. Anon.: Resumé of High Temperature Investigations Conducted during 1945. The Timken Roller Bearing Co., Steel and Tube Div., (Canton, Ohio).

3. Anon.: Resumé of High Temperature Investigations Conducted during 1946-47. The Timken Roller Bearing Co., Steel and Tube Div., (Canton, Ohio).

TABLE I

CHEMICAL COMPOSITION OF SHEET ALLOYS TESTED AT 1700° AND 1800° F

Alloy	Heat number	Chemical Composition (percent)													
		C	Mn	Si	P	S	Cr	Ni	Co	Mo	W	Cb	Fe	N_2	B
Vitallium	------	0.20	----	----	-----	-----	27	2.5	Bal.	5.5	----	----	----	-----	----
Co-Cr-Ni	------	Low	----	----	-----	-----	30	15	55	----	----	----	----	-----	----
S816	------	.35	----	----	-----	-----	18	20	45	4	4	3	----	-----	----
S590	------	.41	0.75	0.55	-----	-----	19.22	18.87	20.56	4.26	4.12	3.70	Bal.	-----	----
Inconel	------	.13	1	.5	-----	-----	13	75	-----	----	----	----	9	-----	----
Low-carbon N-155	------	.14	1.48	.58	-----	-----	21.24	19.53	20.22	3.10	1.95	1.01	Bal.	0.13	----
J-837	J-837	.13	1.69	.50	-----	-----	20.27	4.40	20.82	4.35	4.56	----	Bal.	----	0.34
J-838	J-838	.11	1.46	.51	-----	-----	20.30	14.90	21.14	4.35	4.43	----	Bal.	----	.37
J-839	J-839	.12	1.43	.49	-----	-----	21.70	4.40	20.04	4.48	4.40	----	Bal.	.124	.56
J-840	J-840	.13	1.38	.58	-----	-----	21.29	15.28	20.98	4.33	4.33	----	Bal.	.117	.45
Type 330	8R1445	.18	1.37	.62	0.015	0.018	15.77	33.27	-----	----	----	----	Bal.	-----	----
Type 310	9R826	.14	1.63	1.09	.018	.017	25.50	21.33	-----	----	----	----	Bal.	-----	----
Type 310S (AF-18)	------	.13	----	2	-----	-----	25.	20	-----	----	----	----	Bal.	-----	----
Type 310S	14626	.047	1.92	2.27	.023	.008	24.83	20.49	-----	----	----	----	Bal.	-----	----
Type 309	------	.12	----	----	-----	-----	25	12	-----	----	----	----	Bal.	-----	----

TABLE II

RUPTURE-TEST DATA AT 1700° AND 1800° F FOR 14 SHEET ALLOYS

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)
Vitallium	1700	25,300	[a]S.T.T.T.	13
		21,000	0.50	15
		9,000	56	7
		8,000	103	14
		6,000	209	14
		4,000	599	20
Vitallium	1800	18,900	S.T.T.T.	25
		11,000	2.43	15
		5,000	60	16
		4,000	75	15
		3,500	160	15
		3,000	472	23
		2,750	269	12
Co-Cr-Ni	1700	22,000	S.T.T.T.	6
		13,000	1.13	4
		10,000	16	9
		6,000	52	2
		5,000	140	1.5
		4,000	333.5	3
Co-Cr-Ni	1800	16,000	S.T.T.T.	9
		11,000	1.92	11
		9,000	5.5	8
		8,000	18	13
		4,500	82	5
		4,000	90.5	4
		3,500	192	6
		3,000	132.5	3
		2,500	324	2
S816	1700	26,200	S.T.T.T.	13
		20,000	0.50	12
		14,000	4.67	19
		5,500	64	26
		3,800	113	16
		3,000	256.5	19
		2,500	600	40

[a]Short-time tensile test.

TABLE II - Continued

RUPTURE-TEST DATA AT 1700° AND 1800° F - Continued

Alloy	Temperature (F)	Stress (psi)	Rupture time (hr)	Elongation (percent)
S816	1800	18,000	S.T.T.T	26
		9,000	3.20	30
		3,500	25.5	32
		2,500	83	36
		2,000	136	39
		1,500	259.5	45
S590	1700	21,200	S.T.T.T.	9
		14,000	2.42	13
		11,000	8.08	13
		6,000	118.5	9
		5,000	235	21
		4,000	518	7
S590	1800	15,000	S.T.T.T.	22
		11,000	1.42	17
		8,000	8.6	22
		6,000	21	18
		5,000	28	17
		4,000	110	16
		3,000	180	52
		2,000	693	10
Inconel-1	1700	(b)	S.T.T.T.	(b)
		12,000	0.20	31
		4,000	45.5	19
		3,000	118	27
		2,500	177.5	19
		2,250	1492	96
		2,000	248	36
		1,800	889	57
		2,000	(c)	
Inconel-1	1800	(b)	S.T.T.T	----------
		4,500	5.1	19
		2,500	54	28
		2,000	152.5	25
		1,500	605	44
Inconel-2	1700	4,000	30.5	22
		3,000	175	60
		2,500	649	49
		2,000	1103	50
	1800	2,500	84	35
		2,250	546	39
		2,000	1340	54

[b]Not sufficient material for short-time tensile test.

[c]Overheated at 382 hr.

NACA

TABLE II – Continued

RUPTURE-TEST DATA AT 1700° AND 1800° F – Continued

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)
Low-carbon N-155	1700	(b)	S.T.T.T.	----------
		10,000	21.5	12
		7,000	27	23
		5,000	88	11
		4,500	236	16
		4,000	831	7
Low-carbon N-155	1800	(b)	S.T.T.T.	----------
		6,000	16.5	28
		5,000	17	20
		4,000	81	21
		3,000	176.5	22
		2,500	340	20
J-837	1700	20,400	S.T.T.T.	58
		6,000	33	(d)
		5,000	31	(d)
J-837	1800	12,600	S.T.T.T.	47
		4,000	9	(d)
J-838	1700	20,700	S.T.T.T.	67
		17,000	1.42	54
		13,000	8.5	57
		9,000	33	54
		7,000	132	66
		6,000	284.5	42
		5,000	434	35
J-838	1800	12,900	S.T.T.T.	53
		10,000	2.00	39
		5,000	34.5	29
		4,000	117	61
		3,000	186	40
		2,500	313.5	37
J-839	1700	23,300	S.T.T.T.	62
		6,000	33	(d)
J-839	1800	13,200	S.T.T.T.	33

[b]Not sufficient material for short-time tensile test.
[d]Specimen severely oxidized.

NACA

TABLE II – Continued

RUPTURE-TEST DATA AT 1700° AND 1800° F – Continued

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)
J-840	1700	19,200	S.T.T.T.	73
		14,000	2.67	51
		7,000	46	25
		6,000	160	26
		5,000	144	16
		4,000	214	27
		3,000	1133	15
J-840	1800	14,000	S.T.T.T.	66
		9,000	2.67	40
		4,000	24.5	34
		3,000	52	19
		2,500	232	22
		2,000	233	36
		1,700	1309	28
Type 330	1700	9,740	S.T.T.T.	44
		7,000	2.17	56
		4,000	17	28
		2,700	46.5	36
		2,000	88	25
		2,000	147	34
		1,750	153	43
		1,650	233	33
		1,500	811	60
		1,500	525	60
Type 310	1700	13,700	S.T.T.T.	18
		9,000	2.0	20
		4,000	85	20
		3,000	285.5	15
		2,500	728	10
Type 310	1800	9,650	S.T.T.T.	25
Type 310S (AF-18)	1700	15,600	S.T.T.T.	42
		11,000	1.05	12
		9,000	5.00	5
		5,000	87	7
		4,000	276	6
		3,000	1124	3.5

NACA

TABLE II – Concluded

RUPTURE–TEST DATA AT 1700° AND 1800° F – Concluded

Alloy	Temperature (°F)	Stress (psi)	Rupture time (hr)	Elongation (percent)
Type 310S (AF–18)	1800	12,900	S.T.T.T.	26
		8,000	2.0	14
		6,500	11.2	16
		4,000	75.0	7
		3,500	112	5
		3,000	231	5
		2,500	289	6
		2,200	384	4
Type 310S (14626)	1700	10,200	S.T.T.T.	43
		5,000	25	6
		4,000	64	10
		3,000	233	9
		2,400	397	7
Type 310S (14626)	1800	7,550	S.T.T.T.	33
		4,000	23	54
		3,000	85	8
		2,000	345	4
		1,700	569	5
Type 309	1700	10,390	S.T.T.T.	51
		7,000	2.87	45
		4,000	36	27
		3,000	109	21
		2,500	277	7
Type 309	1800	6,810	S.T.T.T.	51
		4,500	7.5	15
		2,500	95.5	15
		2,300	254	14
		2,000	93.5	18
		1,800	321	10
		1,500	Discontinued after 1772 hr	8

NACA

TABLE III

RUPTURE STRENGTHS OF SHEET ALLOYS AT 1700° AND 1800° F

[From curves of stress against rupture time, fig. 1]

Alloy	Temperature (°F)	Stress for rupture in indicated time periods (psi)				
		1 hr	10 hr	100 hr	400 hr	1000 hr
Vitallium	1700	18,500	12,000	8000	4600	3300
Co-Cr-Ni	1700	14,000	10,500	5600	3700	[a]2800
S816	1700	18,000	10,500	4500	2700	1900
S590	1700	17,000	10,500	6300	4300	3400
Inconel-1	1700	10,000	5,600	3100	2200	1750
Inconel-2	1700	------	4,800	3300	2700	2100
Low-carbon N-155	1700	------	------	5300	4400	3900
J-837	1700	(b)	(b)	(b)	(b)	(b)
J-838	1700	19,000	12,000	7400	5200	[a]3500
J-839	1700	(b)	(b)	(b)	(b)	(b)
J-840	1700	18,000	10,000	5600	3900	3100
Type 330	1700	8,500	4,600	1950	1600	1400
Type 310	1700	10,500	6,400	3800	2850	2300
Type 310S (AF-18)	1700	11,000	8,000	5000	3700	3100
Type 310S (14626)	1700	------	6,400	3600	2500	[a]2000
Type 309	1700	8,800	5,200	3100	2300	[a]1900
Vitallium	1800	13,500	7,800	4200	2500	[a]1800
Co-Cr-Ni	1800	11,500	8,600	4000	2300	[a]1600
S816	1800	11,200	5,600	2200	1270	[a]800
S590	1800	11,800	7,600	3700	2400	1800
Inconel-1	1800	[a]6,500	3,800	2200	1600	1300
Inconel-2	1800	------	-----	2500	2200	2000
Low-carbon N-155	1800	------	6,800	3500	2400	[a]1850
J-838	1800	11,000	6,800	4100	2300	[a]1500
J-840	1800	13,000	5,500	2700	2100	1800
Type 310S (AF-18)	1800	9,000	6,600	3600	2200	[a]1600
Type 310S (14626)	1800	------	4,800	2900	1900	1400
Type 309	1800	[a]7,000	4,100	2450	1800	1500

NACA

[a]Obtained by extrapolation.
[b]Low because of severe oxidation.

TABLE IV

ESTIMATED RUPTURE ELONGATIONS OF SHEET ALLOYS AT 1700° AND 1800° F

Alloy	Temperature (°F)	Estimated elongation to fracture in indicated time periods (percent)				
		1 hr	10 hr	100 hr	400 hr	1000 hr
Vitallium	1700	15	15	14	15	20
Co-Cr-Ni	1700	4	9	2	3	--
S816	1700	15	20	20	20	40
S590	1700	13	13	9	15	7
Inconel-1	1700	30	25	25	40	55
Inconel-2	1700	--	20	40	55	50
Low-carbon N-155	1700	--	12	11	15	7
J-838	1700	60	57	60	40	--
J-840	1700	55	50	25	20	15
Type 330	1700	55	30	25	40	60
Type 310	1700	20	20	20	15	10
Type 310S (AF-18)	1700	12	5	7	6	4
Type 310S (14626)	1700	--	6	10	7	--
Type 309	1700	50	40	21	7	--
Vitallium	1800	20	15	15	12	--
Co-Cr-Ni	1800	10	10	4	2	--
S816	1800	30	30	35	45	--
S590	1800	17	22	16	30	10
Inconel-1	1800	--	19	26	30	45
Inconel-2	1800	--	--	35	40	50
Low-carbon N-155	1800	--	28	21	20	--
J-838	1800	45	30	60	37	--
J-840	1800	45	40	20	25	28
Type 310S (AF-18)	1800	15	15	6	4	--
Type 310S (14626)	1800	--	55	8	4	--
Type 309	1800	50	15	15	10	--

NACA

TABLE V

COMPARATIVE 1700° AND 1800° F RUPTURE PROPERTIES OF SEVEN ALLOYS

IN THE FORM OF SHEET, BAR STOCK, AND PRECISION CASTINGS

Alloy	Temperature (°F)	Form	Treatment (a)	Rupture strength (psi)		Estimated rupture ductility (percent)		Source of data, reference
				100 hr	1000 hr	100 hr	1000 hr	
Vitallium	1700	Sheet	Annealed	8,000	3,300	14	20	
Vitallium	1700	Cast	--------	13,000	10,000	8	5	1
Vitallium	1800	Sheet	Annealed	4,200	[b]1,800	15	--------	
Vitallium	1800	Cast	--------	9,400	7,000	13	12	1
S816	1700	Sheet	A.C.2200–2250° F	4,500	1,900	20	40	
S816	1700	Bar stock	W.Q.2300° F+ 16 hr at 1700° F	9,500	6,100	15	15	1
S816	1700	Cast	----------------	14,500	11,500	10	6	1
S816	1800	Sheet	A.C.2200–2250° F	2,200	[b]800	35	--------	
S816	1800	Bar stock	W.Q.2300° F+ 16 hr at 1800° F	5,300	3,000	10	--------	1
S816	1800	Cast	----------------	10,500	7,800	10	--------	1
S590	1700	Sheet	A.C.2150° F	6,300	3,400	9	7	1
S590	1700	Bar stock	W.Q.2270° F+ 16 hr at 1700° F	9,400	6,600	30	14	1
S590	1700	Cast	----------------	11,000	8,400	15	8	1
S590	1800	Sheet	A.C.2150° F	3,700	1,800	16	10	
S590	1800	Bar stock	W.Q.2270° F+ 16 hr at 1800° F	5,600	3,500	25	--------	1
S590	1800	Cast	----------------	8,000	5,800	15	5	1
Low-carbon N-155	1700	Sheet	A.C.2000–2100° F	5,300	3,900	11	7	
Low-carbon N-155	1700	Bar stock	Hot-rolled	5,100	2,500	30	--------	1
Low-carbon N-155	1700	Bar stock	A.C.2282° F	7,600	4,800	20	10	1
Low-carbon N-155	1800	Sheet	A.C.2000–2100° F	3,500	[b]1,850	21	--------	
Low-carbon N-155	1800	Bar stock	A.C.2282° F	4,900	2,800	11	--------	1
25-12 (309)	1700	Sheet	Annealed	3,100	[b]1,900	21	--------	
25-12	1700	Bar stock	W.Q.2200° F	4,000	2,100			2
25-12	1700	Bar stock	W.Q.1900° F	2,700	1,300			2
25-12 (309)	1800	Sheet	Annealed	2,450	1,500	15	--------	
25-12	1800	Bar stock	W.Q.2200° F	2,200	1,050			2
25-12	1800	Bar stock	W.Q.1900° F	1,800	1,000			2
35-15 (330)	1700	Sheet	A.C.1950° F	1,950	1,400	25	60	
35-15	1700	Bar stock	W.Q.2000° F	3,000	1,800			2
35-15	1800	Bar stock	W.Q.2000° F	2,000	1,200			2
25-20 (310)	1700	Sheet	A.C.2050° F	3,800	2,300	20	10	
25-20	1700	Bar stock	W.Q.2150° F	4,200	3,500			2
25-20	1700	Bar stock	N. 1700° F	1,800	1,400			2
25-20 (310)	1800	Bar stock	W.Q.2150 F	2,750	1,900			2
25-20	1800	Bar stock	N. 1700 F	1,500	820			2

[a]A.C. – air-cooled; W.Q. – water-quenched; N. – normalized.
[b]Obtained by extrapolation.

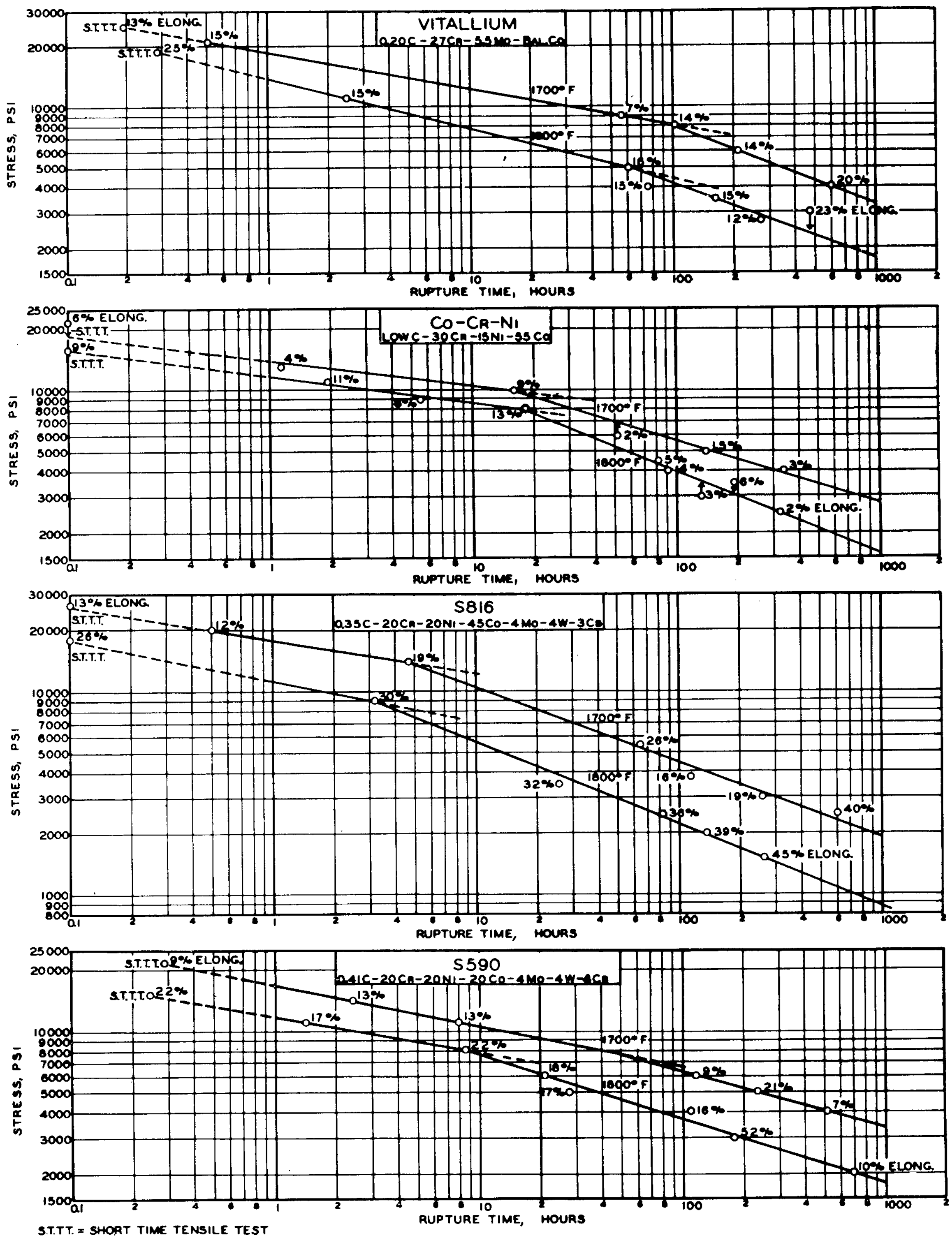
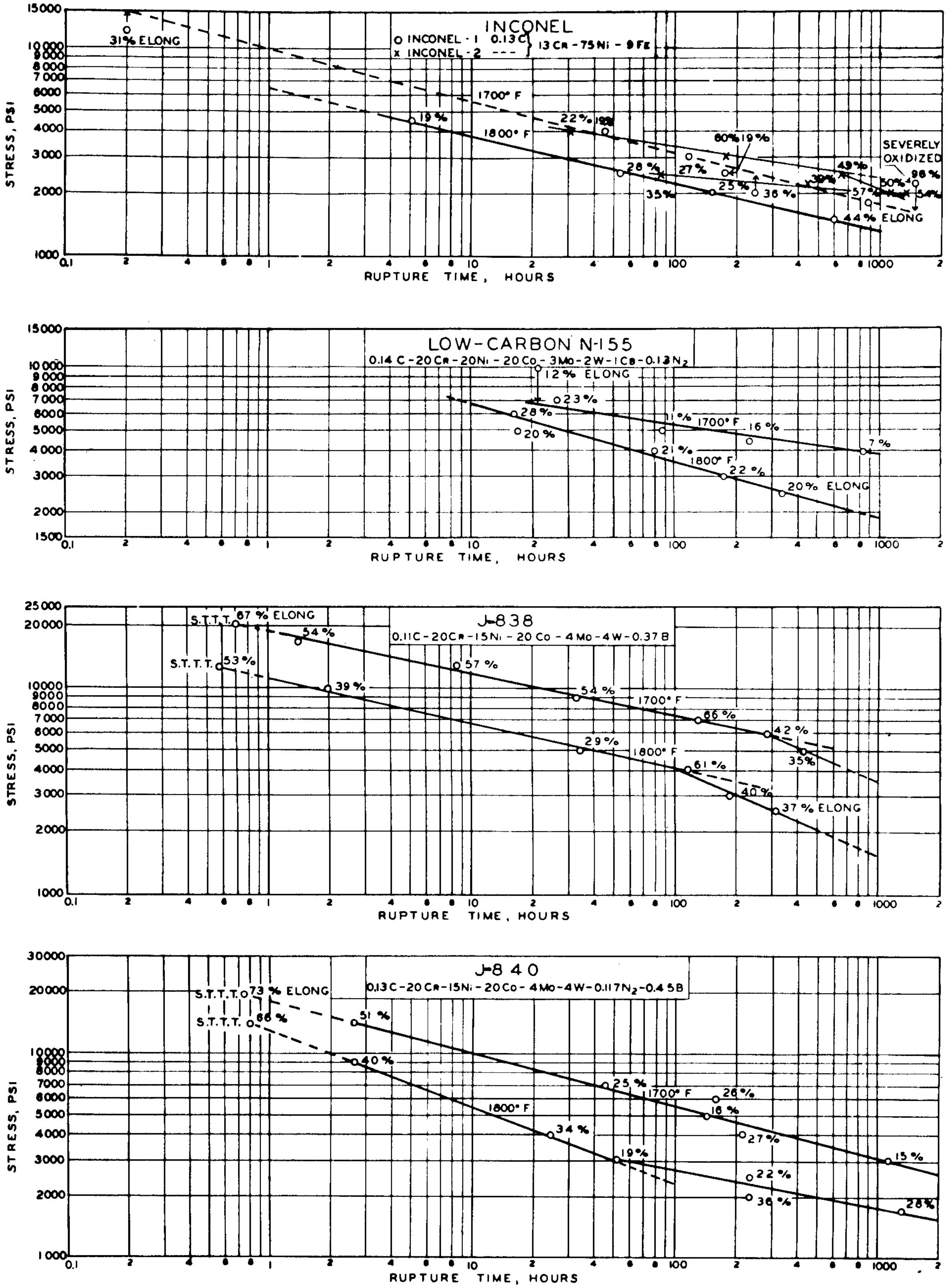


Figure 1.- Stress-rupture-time curves at 1700° and 1800° F for sheet alloys.



S.T.T.T. = SHORT TIME TENSILE TEST

Figure 1.- Continued.

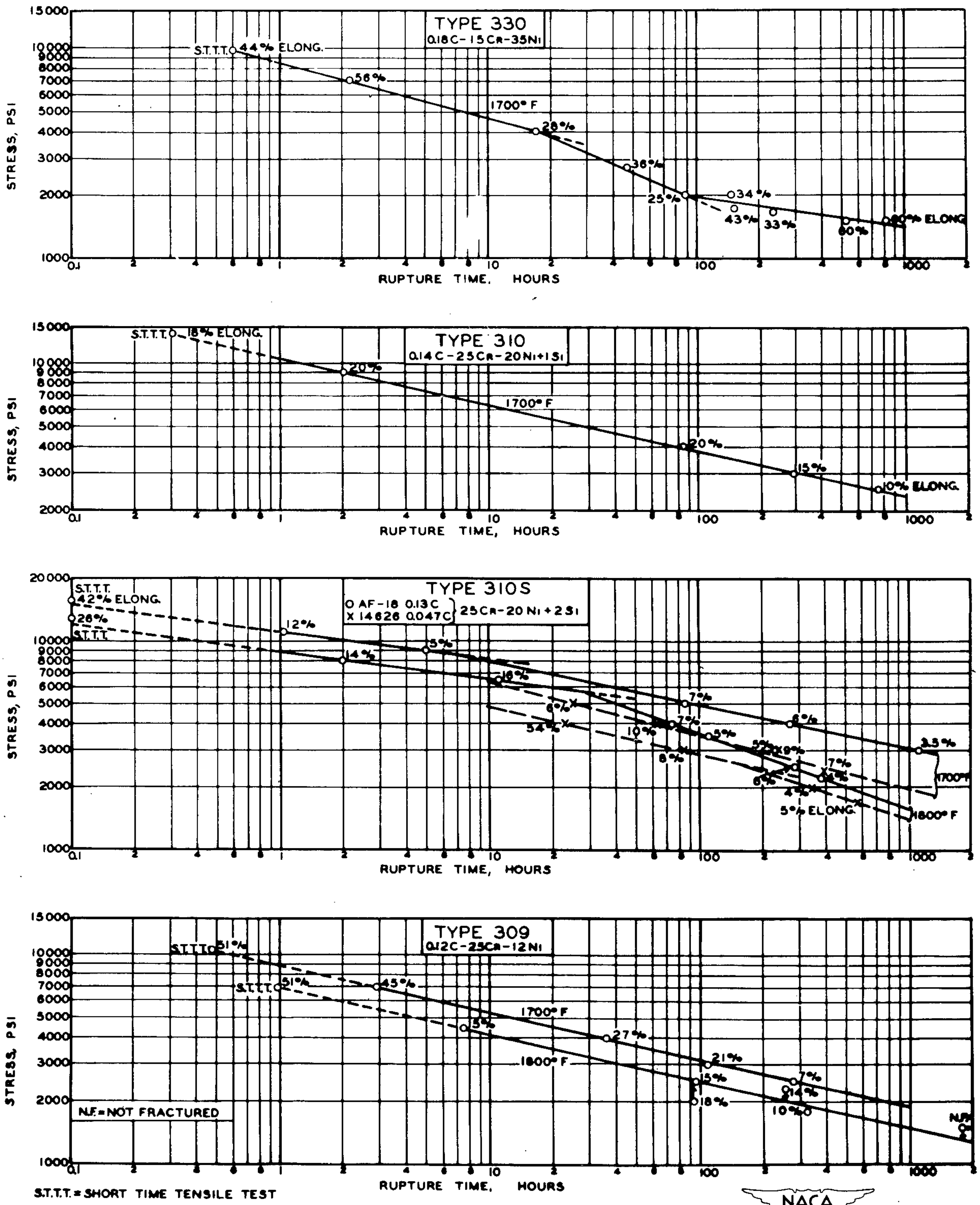
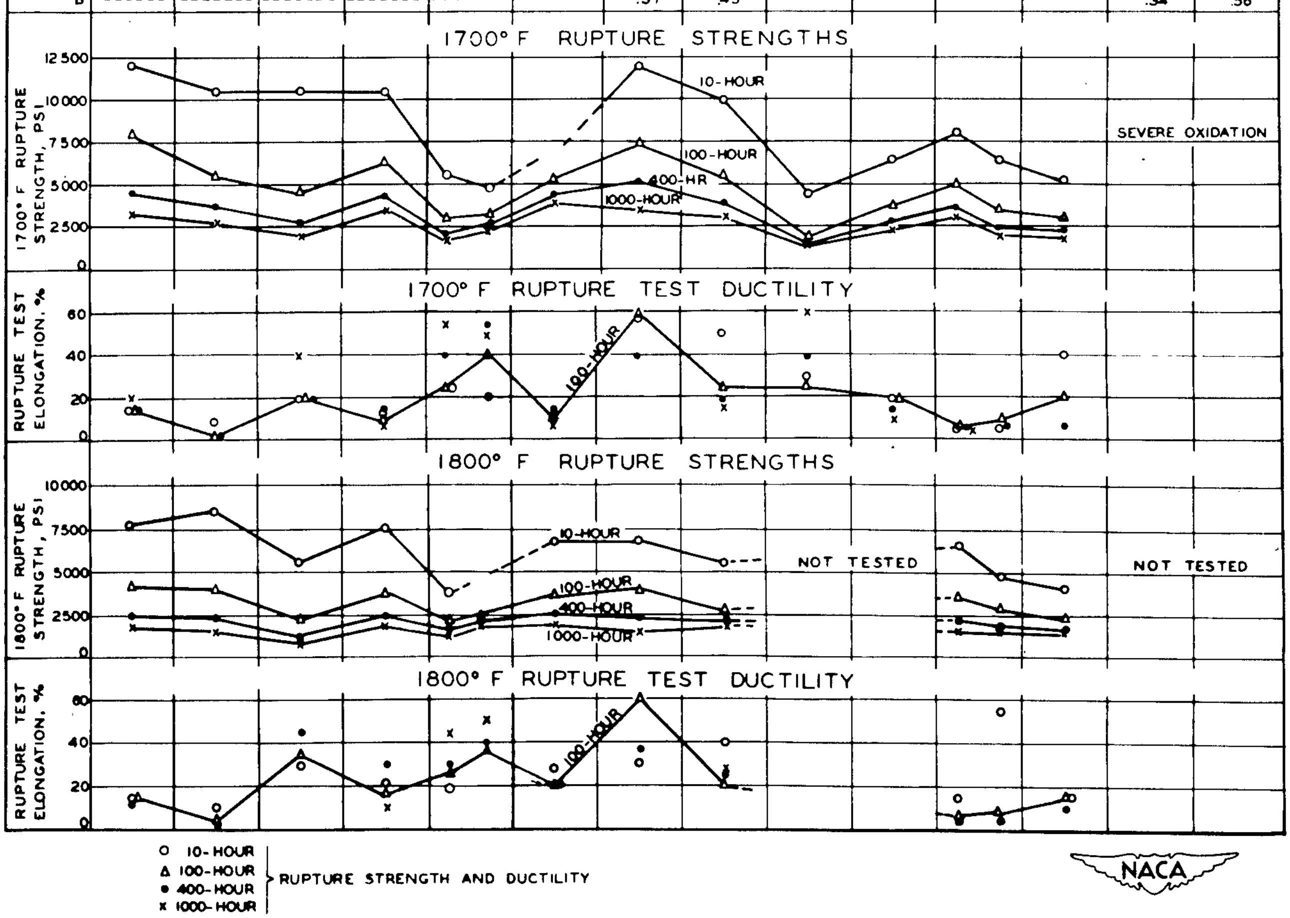


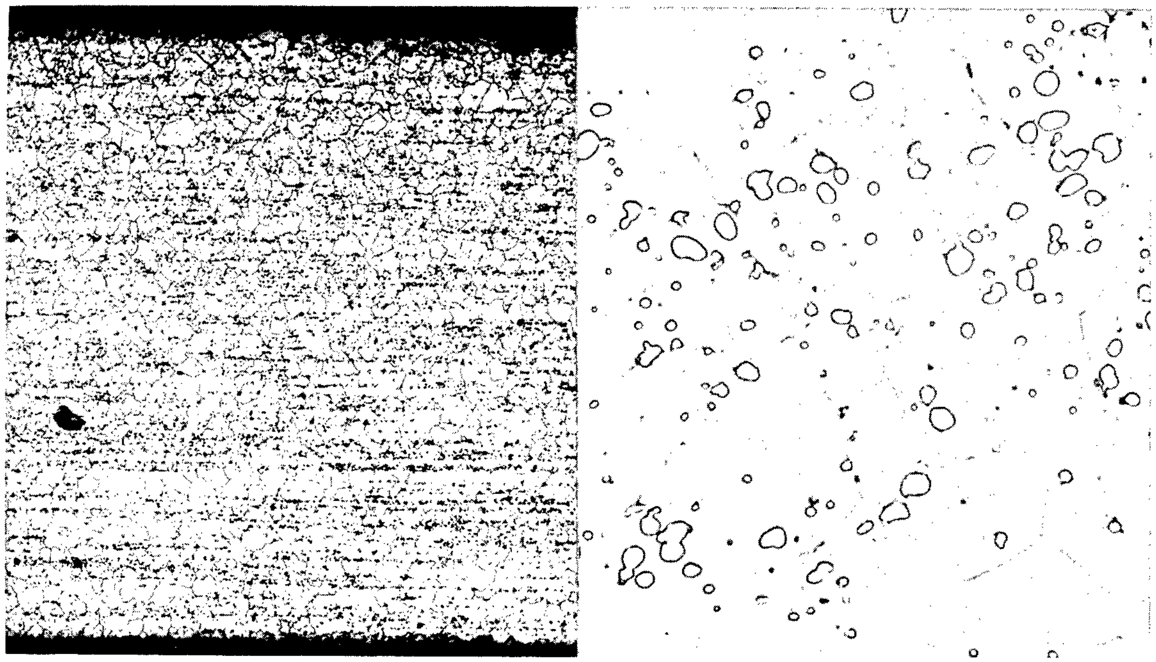
Figure 1.- Concluded

ALLOY	VITALLIUM	Co-Cr-Ni	S816	S590	INCONEL 1 2	LOW-CARBON N155	J-838	J-840	TYPE 330	TYPE 310	TYPE 310S	TYPE 309	J-837	J-839
CHEMICAL COMPOSITION, %														
C	0.20	LOW	0.35	0.41	0.13	0.14	0.11	0.13	0.18	0.14	0.13 0.047	0.12	0.13	0.12
Mn	-------	-------	-------	.75	1.	1.48	1.46	1.38	1.37	1.63	--- 1.92	-------	1.69	1.43
Si	-------	-------	-------	.55	.5	.58	.51	.58	.62	1.09	2. 2.27	-------	.50	.49
Cr	27.	30.	18.	19.22	13.	21.24	20.30	21.29	15.77	25.50	25. 24.83	25.	20.27	21.70
Ni	2.5	15.	20.	18.87	75.	19.53	14.90	15.28	33.27	21.33	20 20.49	12.	4.40	4.40
Co	BAL.	55.	45.	20.56	-------	20.22	21.14	20.98	-------	-------	-------	-------	20.82	20.04
Mo	5.5	-------	4.	4.26	-------	3.10	4.35	4.33	-------	-------	-------	-------	4.35	4.48
W	-------	-------	4.	4.12	-------	1.95	4.43	4.33	-------	-------	-------	-------	4.56	4.40
Cb	-------	-------	3.	3.70	-------	1.01	-------	-------	-------	-------	-------	-------	-------	-------
Fe	-------	-------	-------	-------	9.	-------	-------	-------	-------	-------	-------	-------	-------	-------
N_2	-------	-------	-------	-------	-------	.13	-------	.117	-------	-------	-------	-------	-------	.124
B	-------	-------	-------	-------	-------	-------	.37	.45	-------	-------	-------	-------	.34	.56



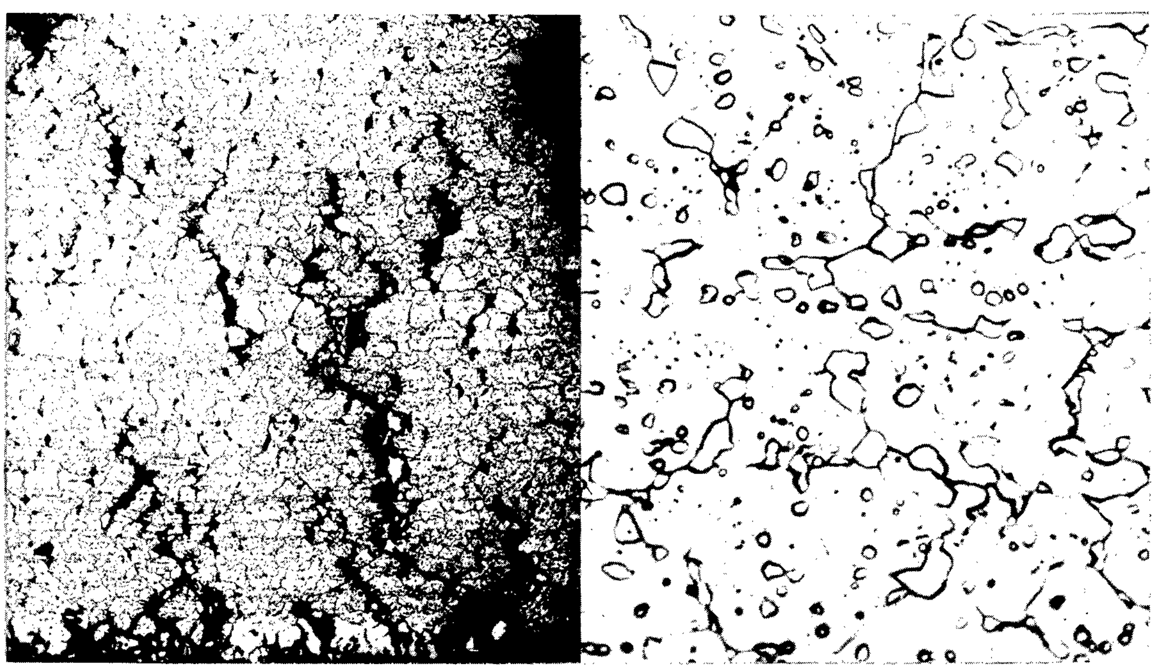
NACA

Figure 2.- Comparative rupture properties at 1700° and 1800° F for sheet alloys.



100X 1000X

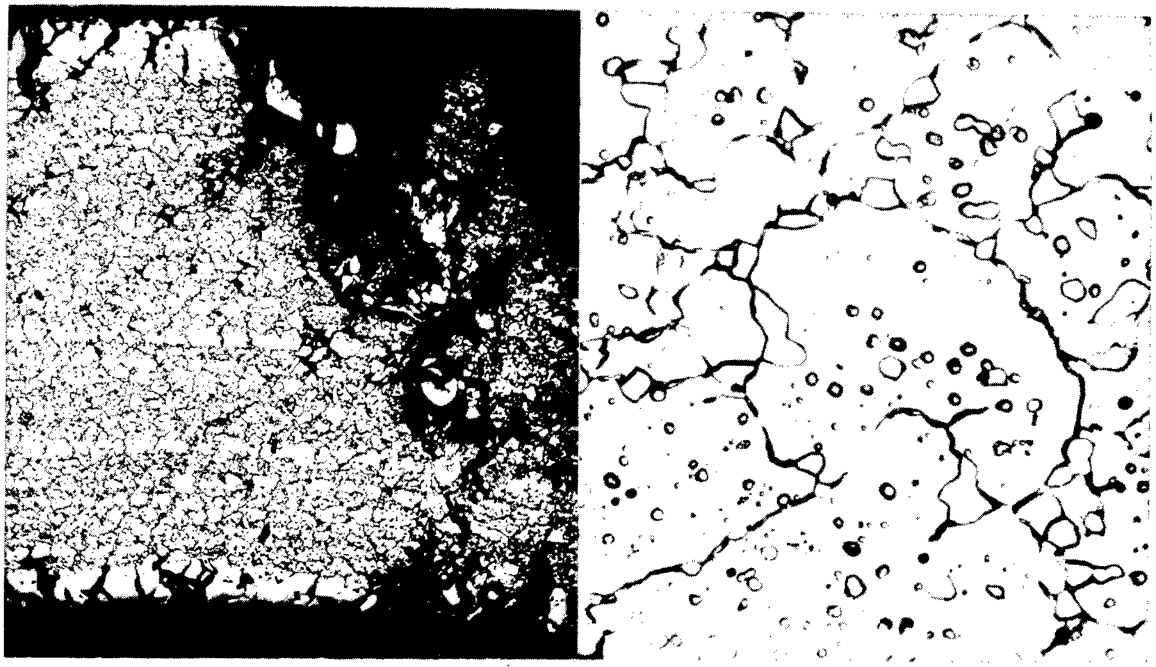
(a) Original microstructure.



Fracture - 100X Interior - 1000X

(b) 599 hours for rupture at 1700° F under 4000 psi.

Figure 3.- Microstructure of sheet Vitallium alloy. Electrolytic chromic acid etch.

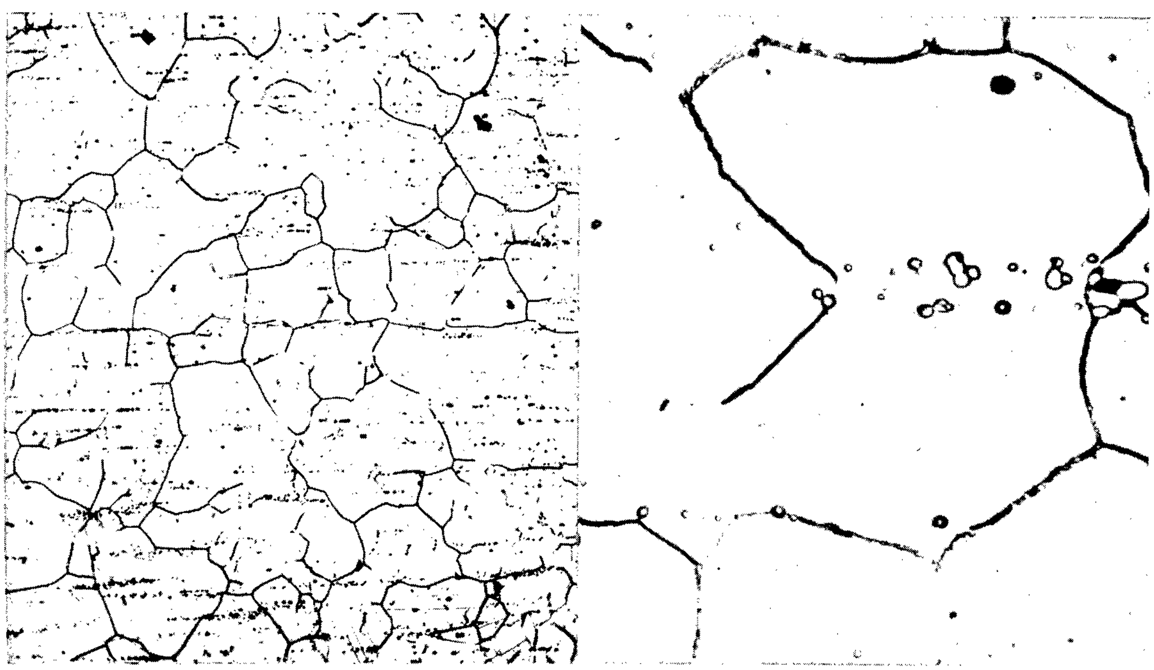


Fracture - 100X

Interior - 1000X

(c) 472 hours for rupture at 1800° F under 3000 psi.

Figure 3.- Concluded.

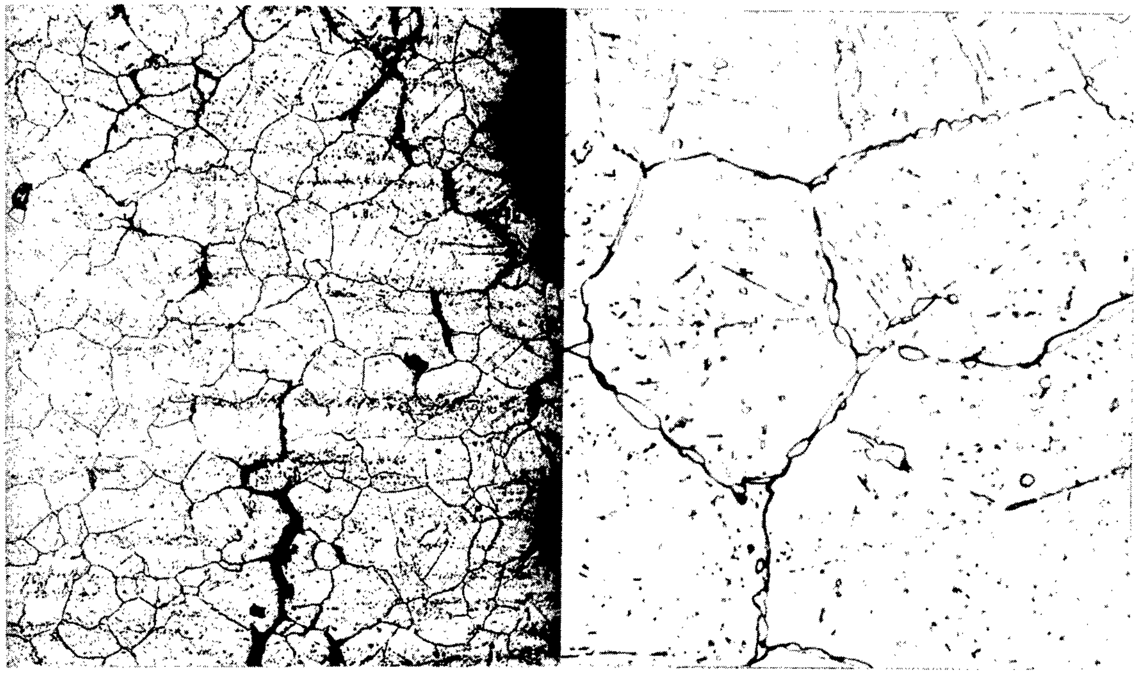


100X

1000X

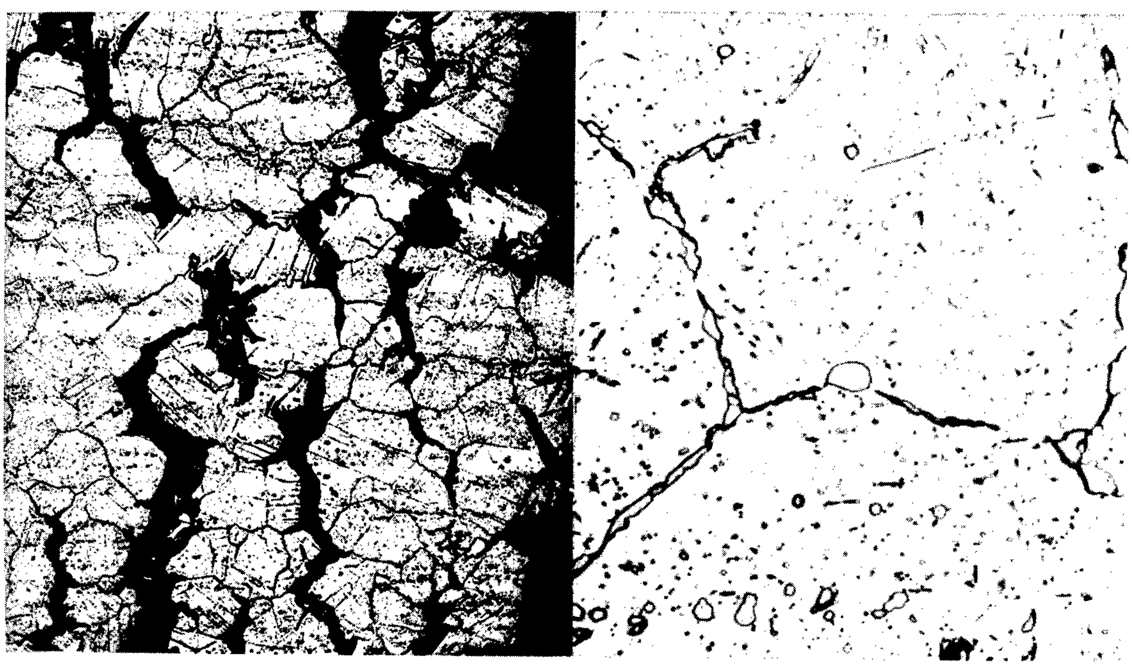
(a) Original microstructure.

Figure 4.- Microstructures of sheet Co-Cr-Ni alloy. Electrolytic chromic acid etch.



Fracture - 100X Interior - 1000X

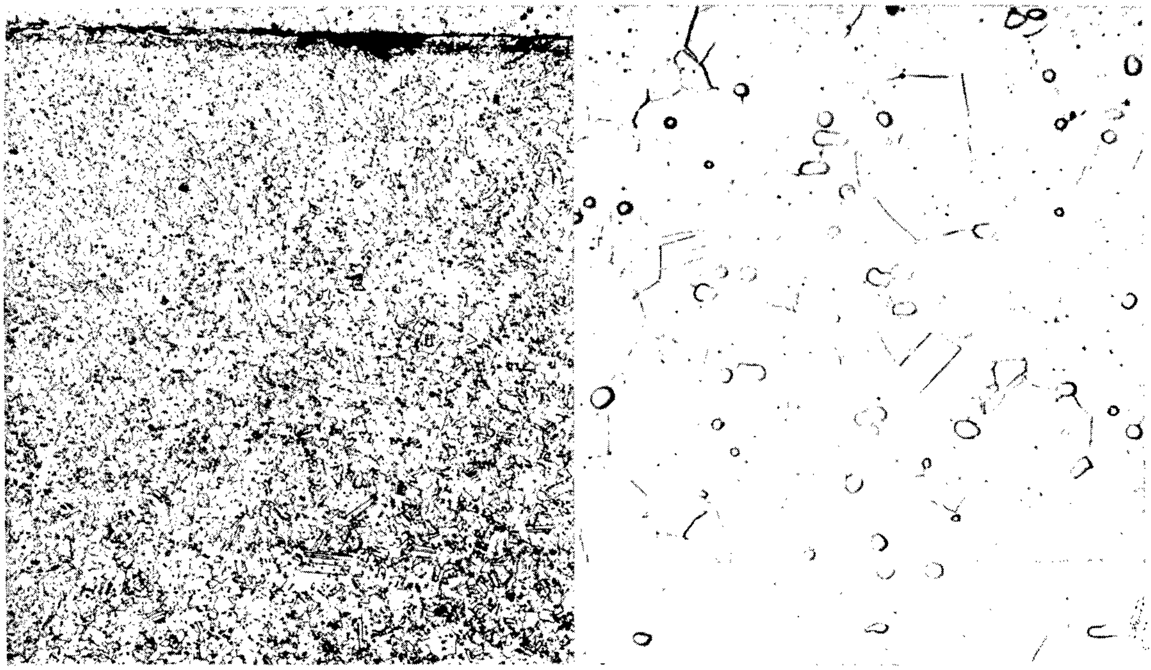
(b) 333.5 hours for rupture at 1700° F under 4000 psi.



Fracture - 100X Interior - 1000X

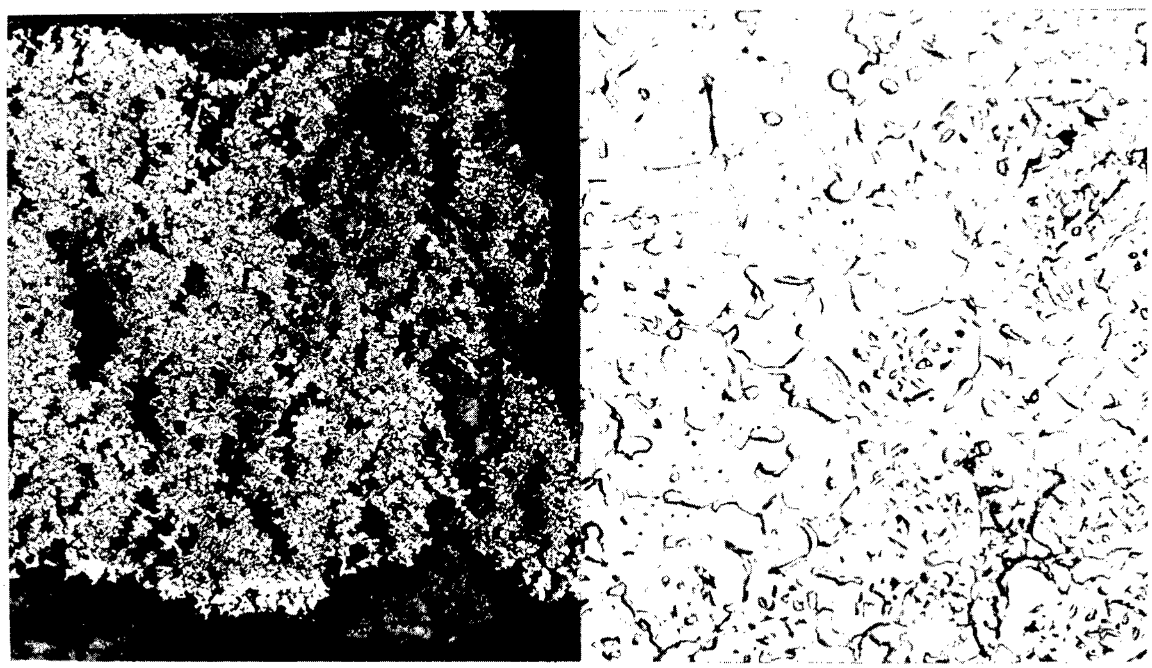
(c) 324 hours for rupture at 1800° F under 2500 psi.

Figure 4.- Concluded



100X 1000X

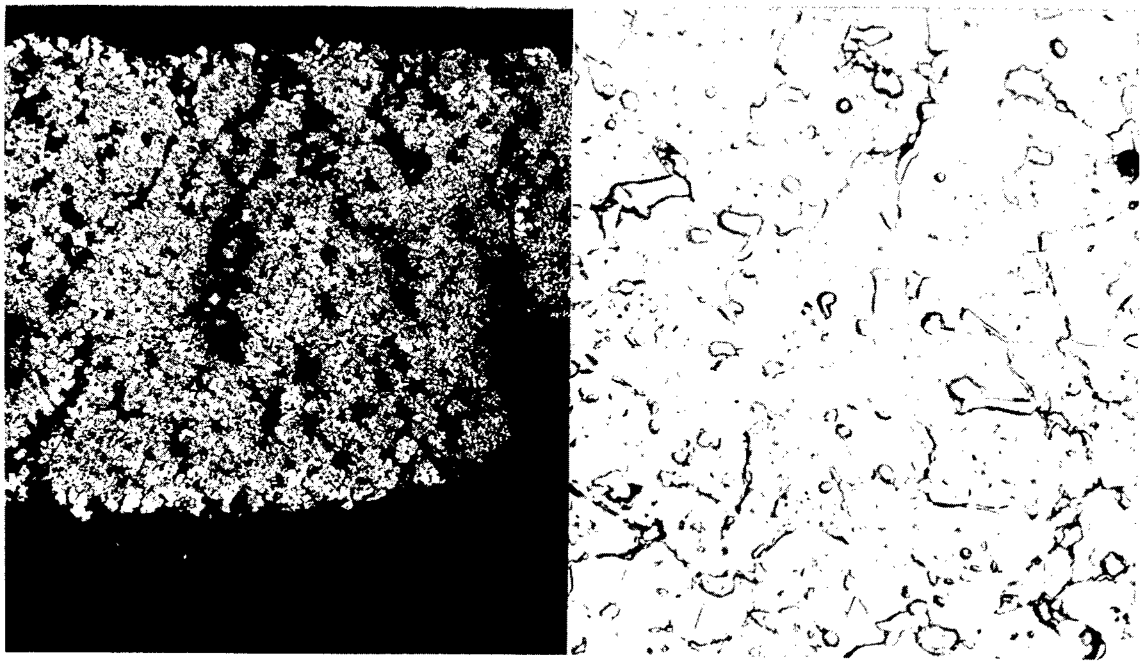
(a) Original microstructure.



Fracture - 100X Interior - 1000X

(b) 600 hours for rupture at 1700° F under 2500 psi.

Figure 5.- Microstructures of sheet S816 alloy. Acid ferric chloride etch.

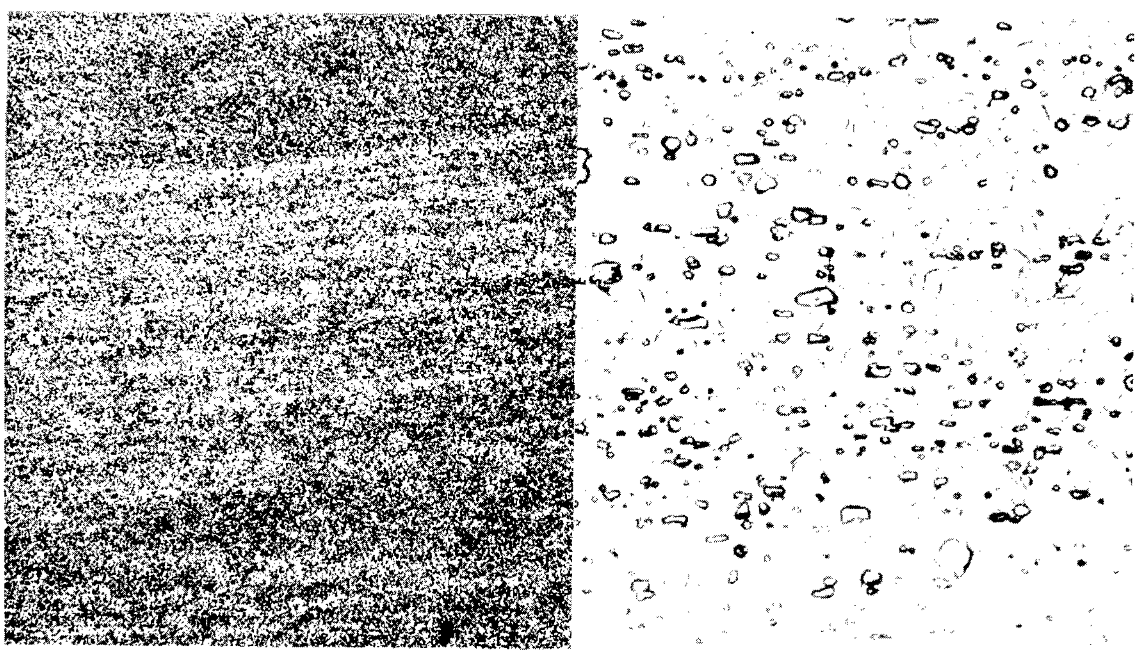


Fracture - 100X

Interior - 1000X

(c) 259.5 hours for rupture at 1800° F under 1500 psi.

Figure 5.- Concluded.

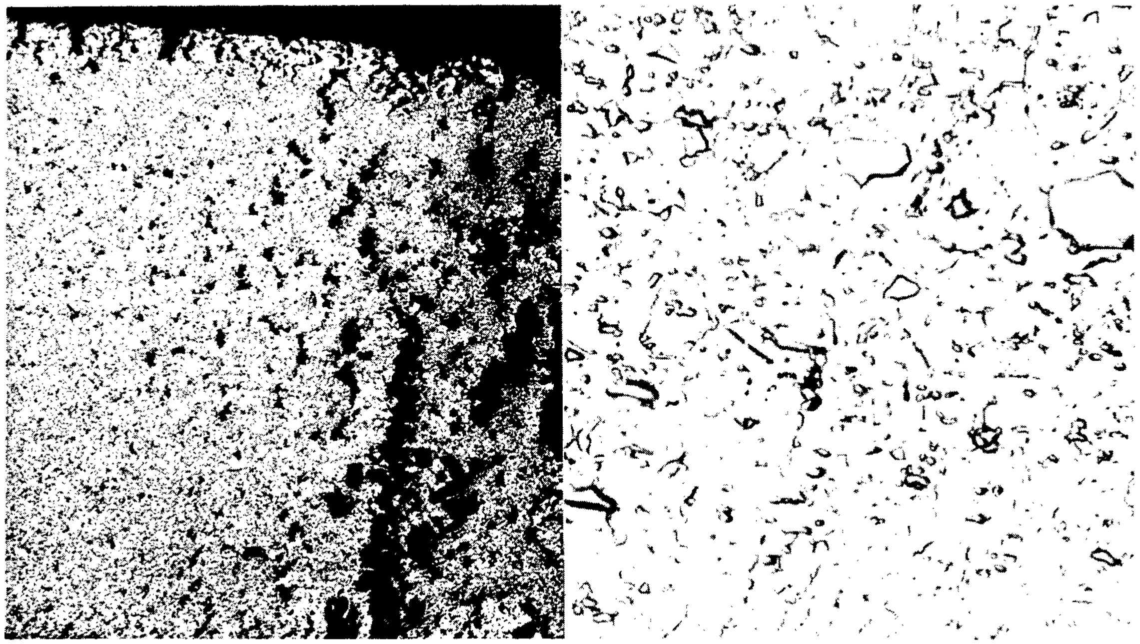


100X

1000X

(a) Original microstructure.

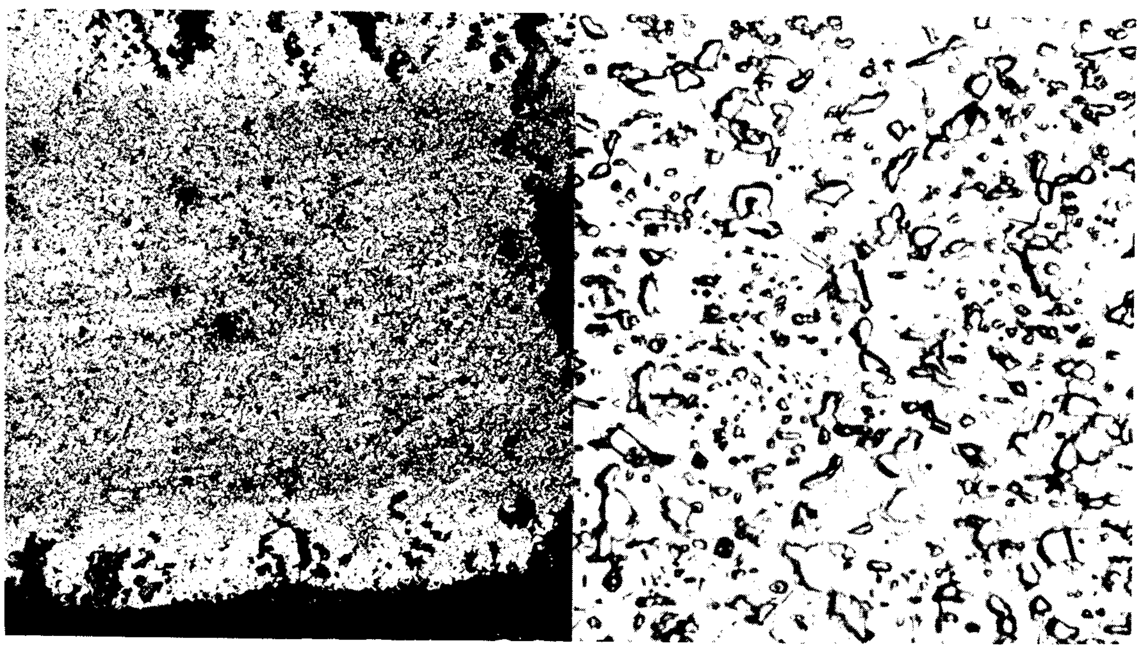
Figure 6.- Microstructures of sheet S590 alloy. Acid ferric chloride etch.



Fracture - 100X

Interior - 1000X

(b) 518 hours for rupture at 1700° F under 4000 psi.

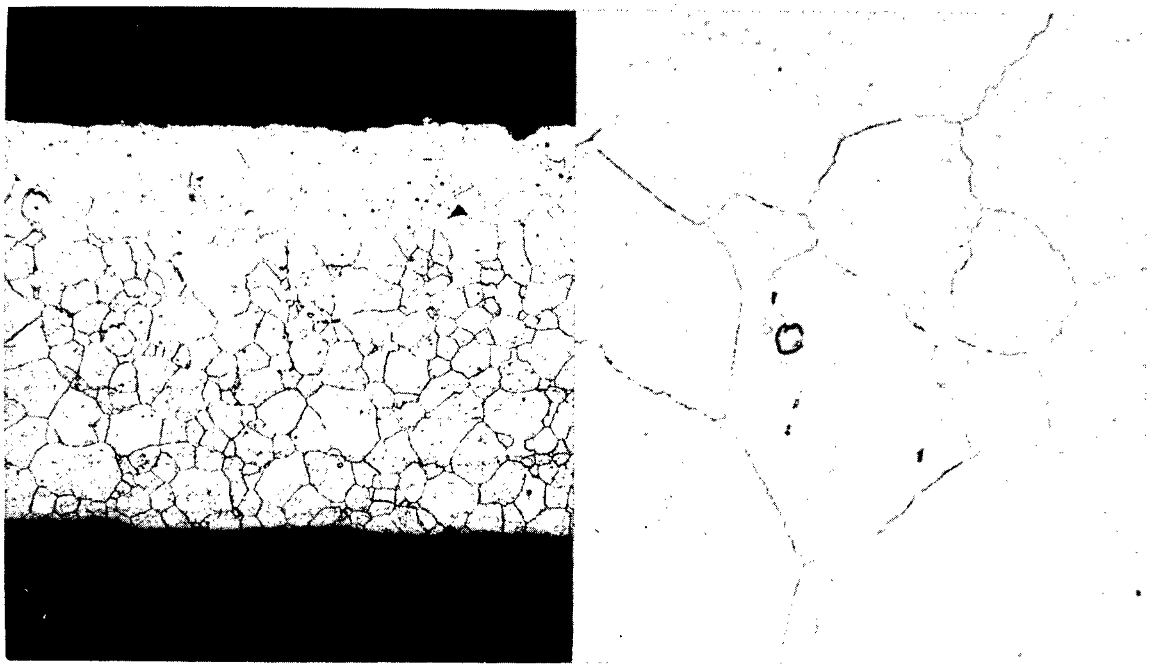


Fracture - 100X

Interior - 1000X

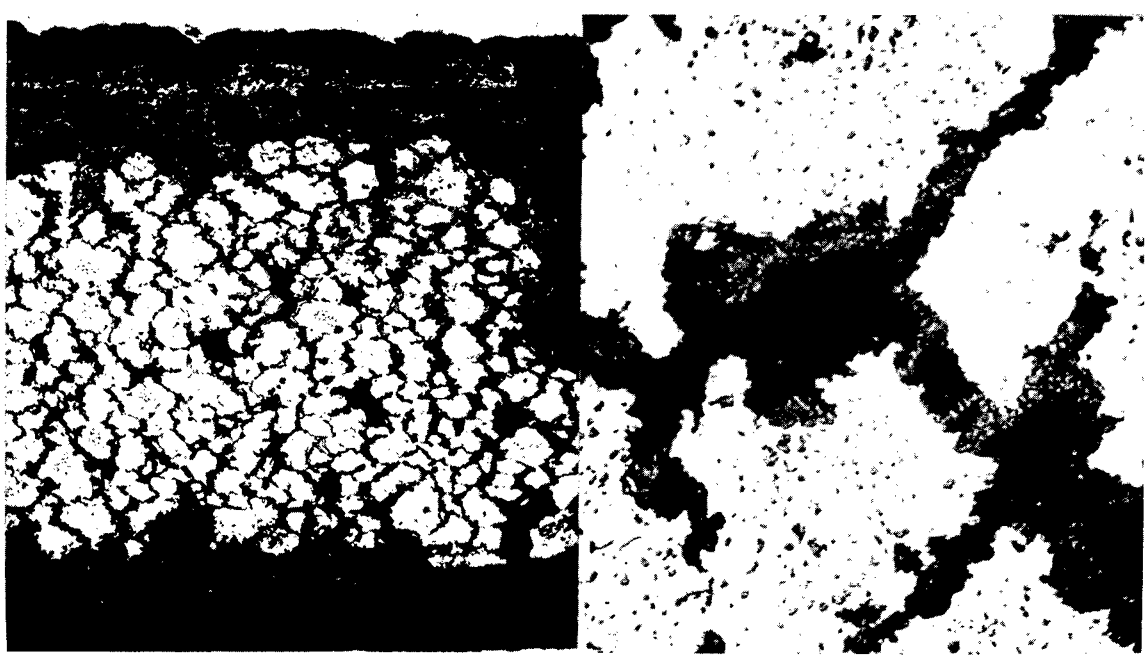
(c) 693 hours for rupture at 1800° F under 2000 psi.

Figure 6.- Concluded.



100X 1000X

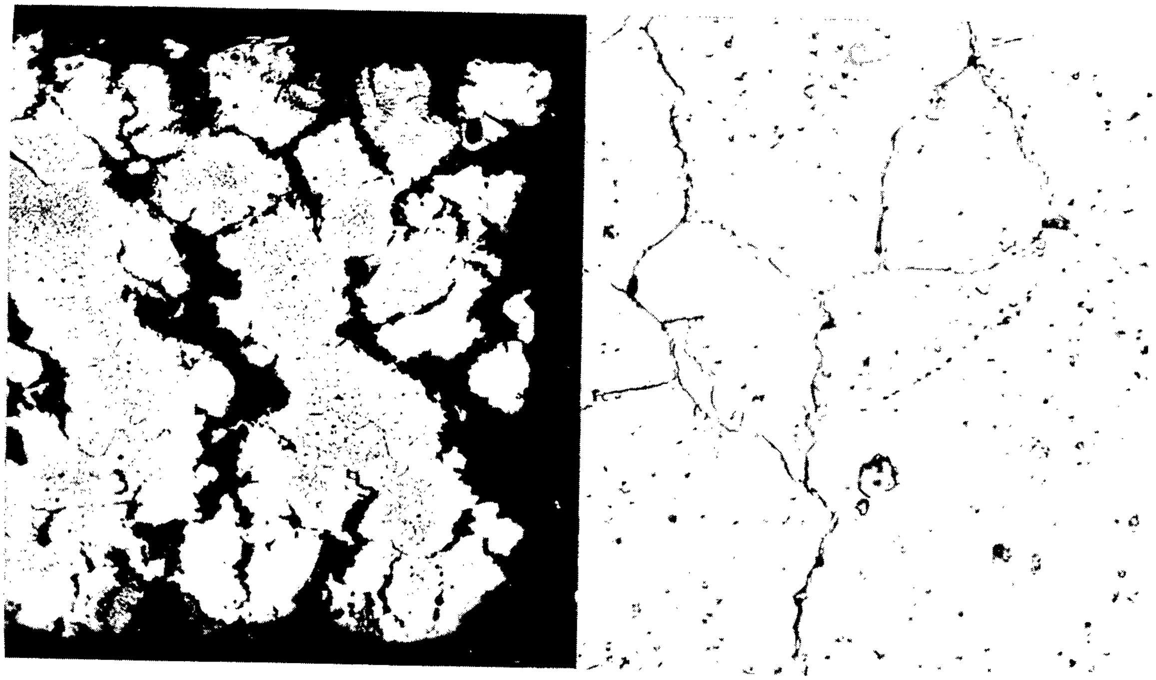
(a) Original microstructure.



Fracture - 100X Interior - 1000X

(b) 1492 hours for rupture at 1700° F under 2250 psi.

Figure 7.- Microstructures of sheet Inconel-1 alloy. Electrolytic chromic acid etch.

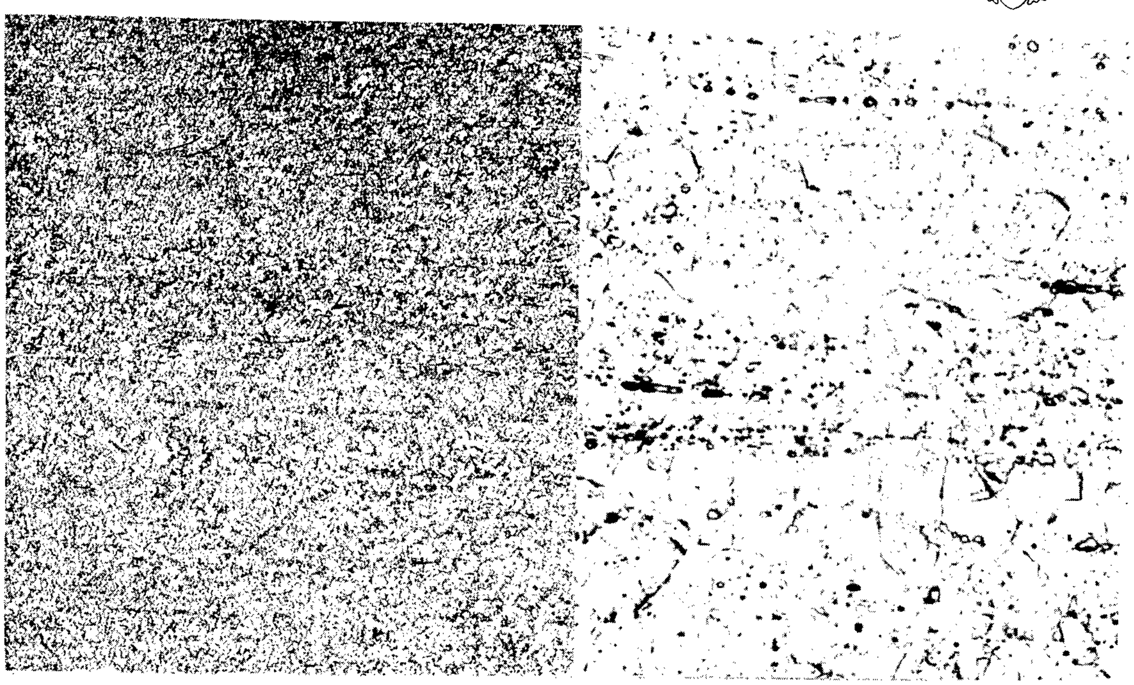


Fracture - 100X

Interior - 1000X

(c) 605 hours for rupture at 1800° F under 1500 psi.

Figure 7.- Concluded.

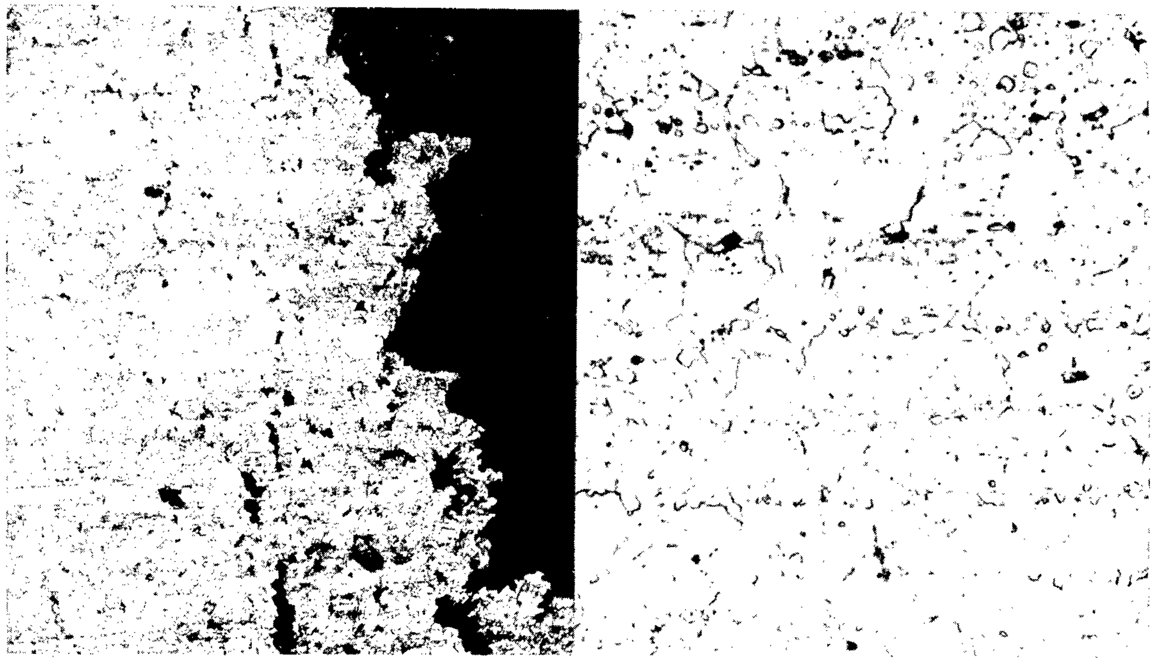


100X

1000X

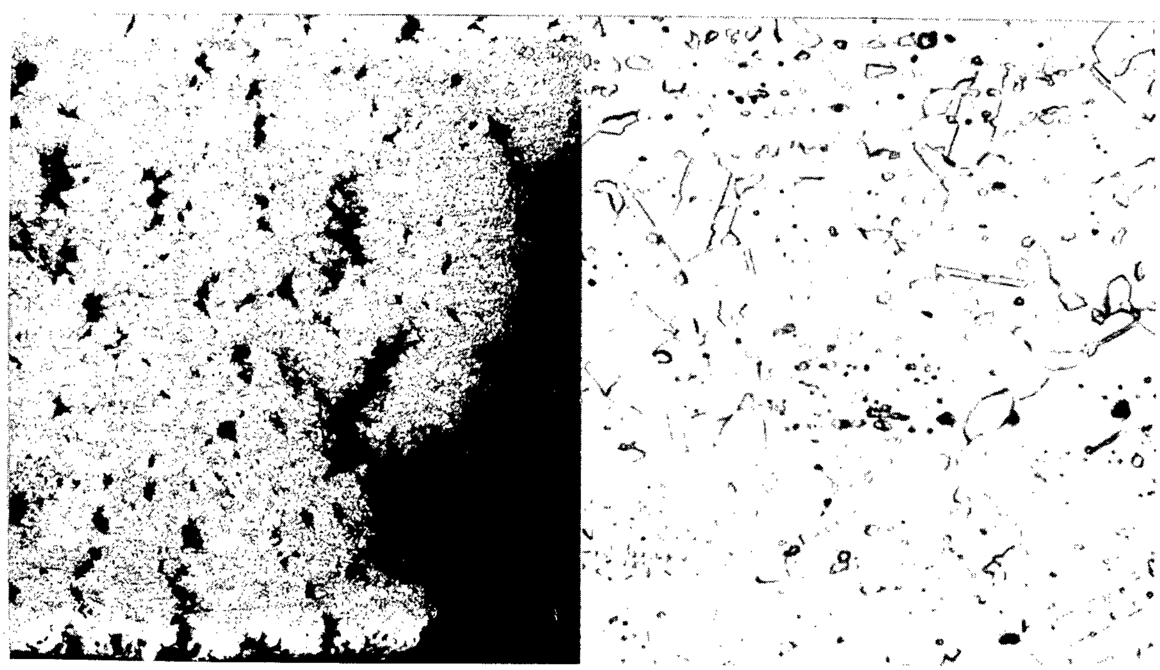
(a) Original microstructure.

Figure 8.- Microstructures of sheet low-carbon N-155 alloy. Acid ferric chloride etch.



Fracture - 100X Interior - 1000X

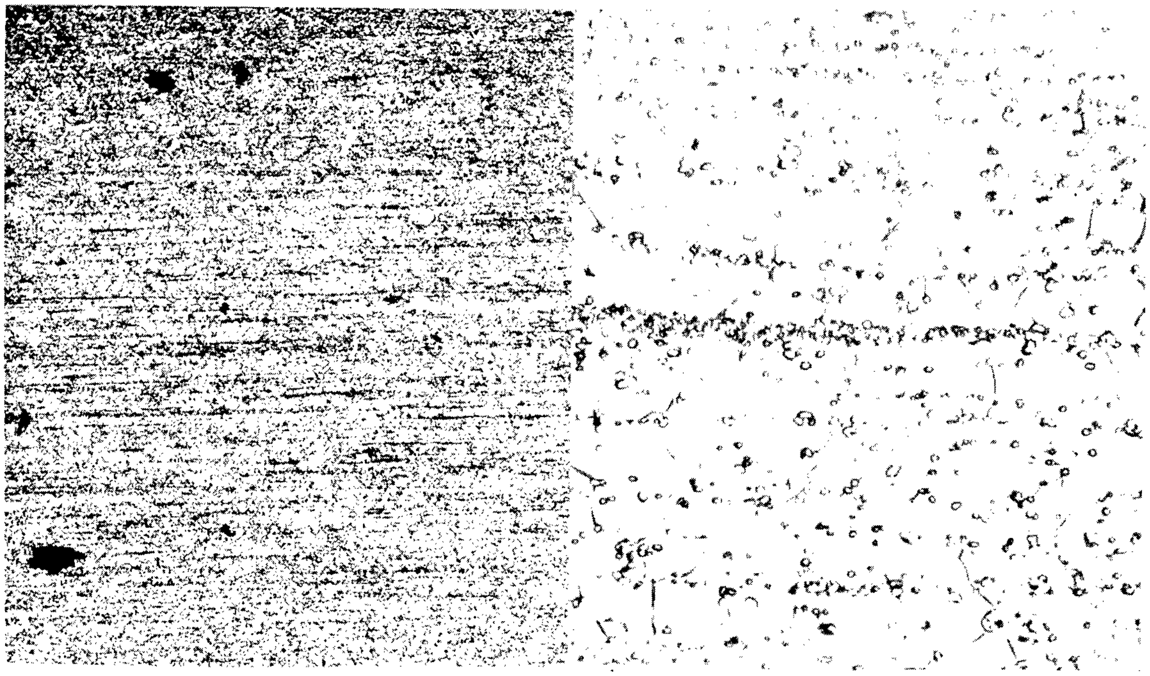
(b) 831 hours for rupture at 1700° F under 4000 psi.



Fracture - 100X Interior - 1000X

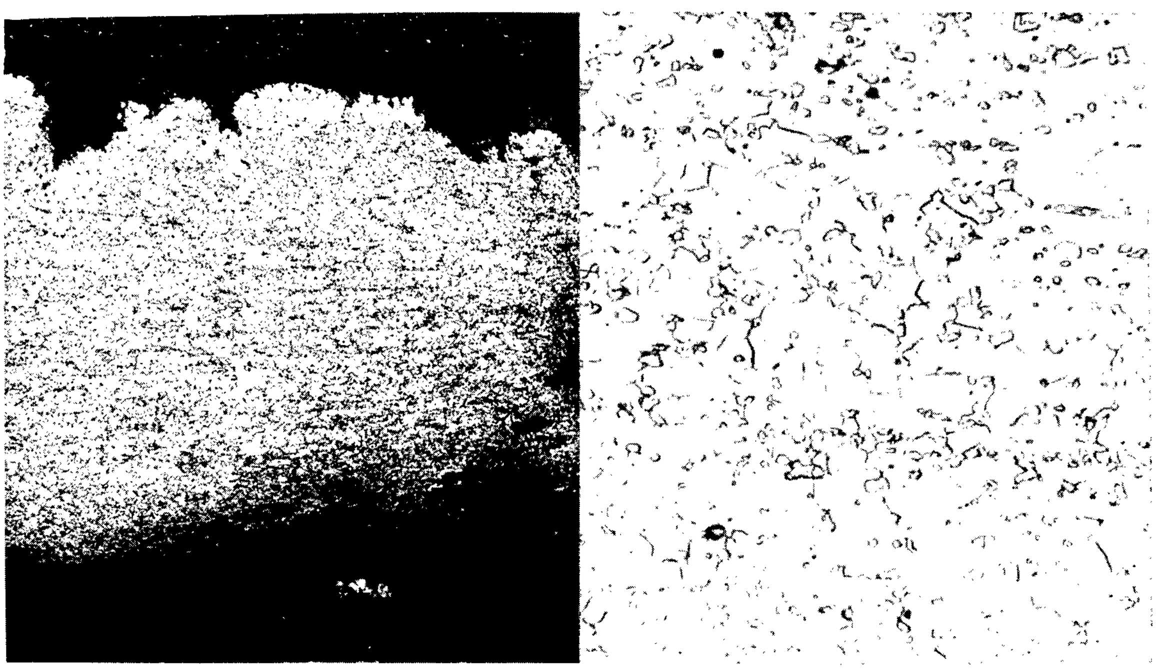
(c) 340 hours for rupture at 1800° F under 2500 psi.

Figure 8.- Concluded.



100X 1000X

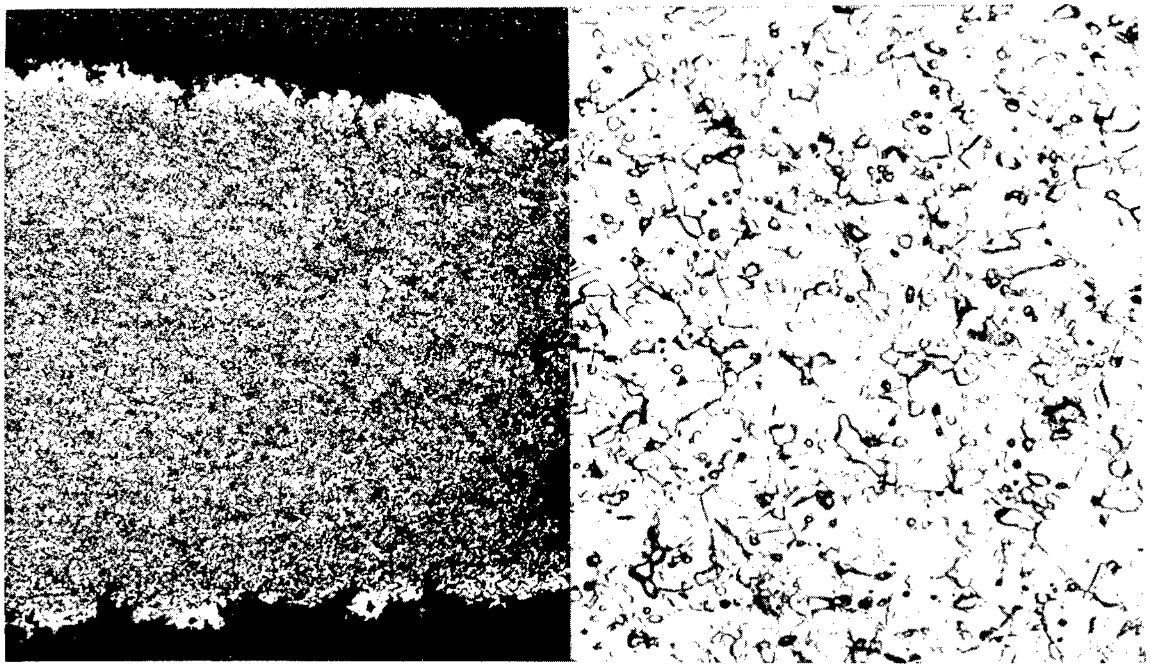
(a) Original microstructure.



Fracture - 100X Interior - 1000X

(b) 434 hours for rupture at 1700° F under 5000 psi.

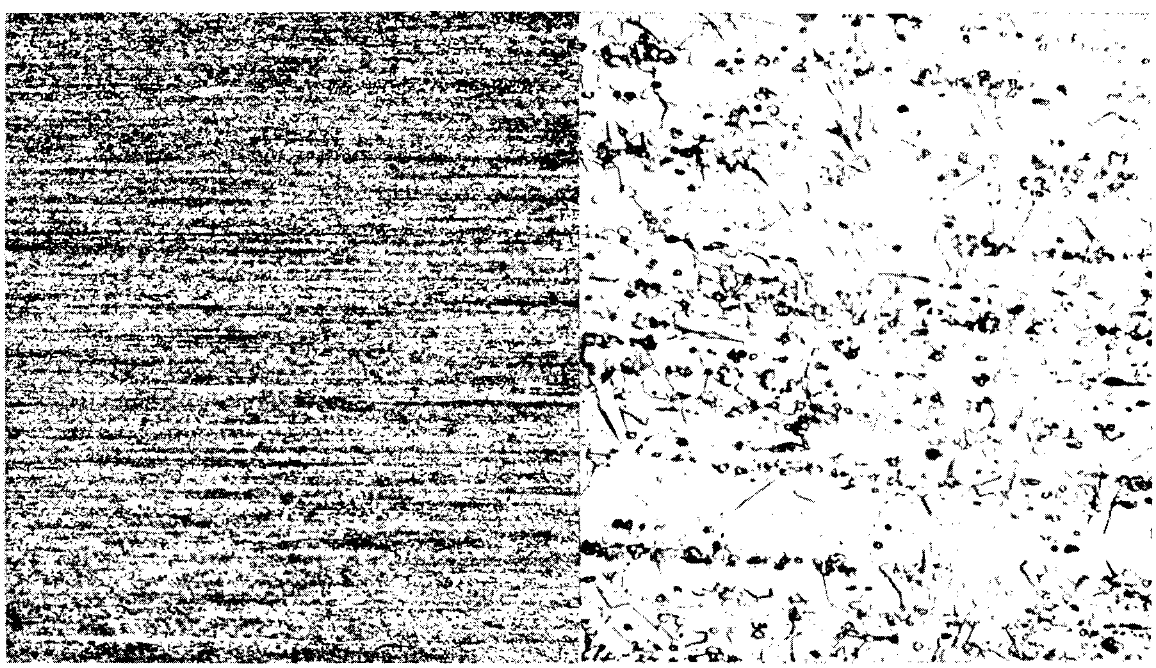
Figure 9.- Microstructures of sheet J-838 alloy. Acid ferric chloride etch.



Fracture - 100X Interior - 1000X

(c) 313.5 hours for rupture at 1800° F under 2500 psi.

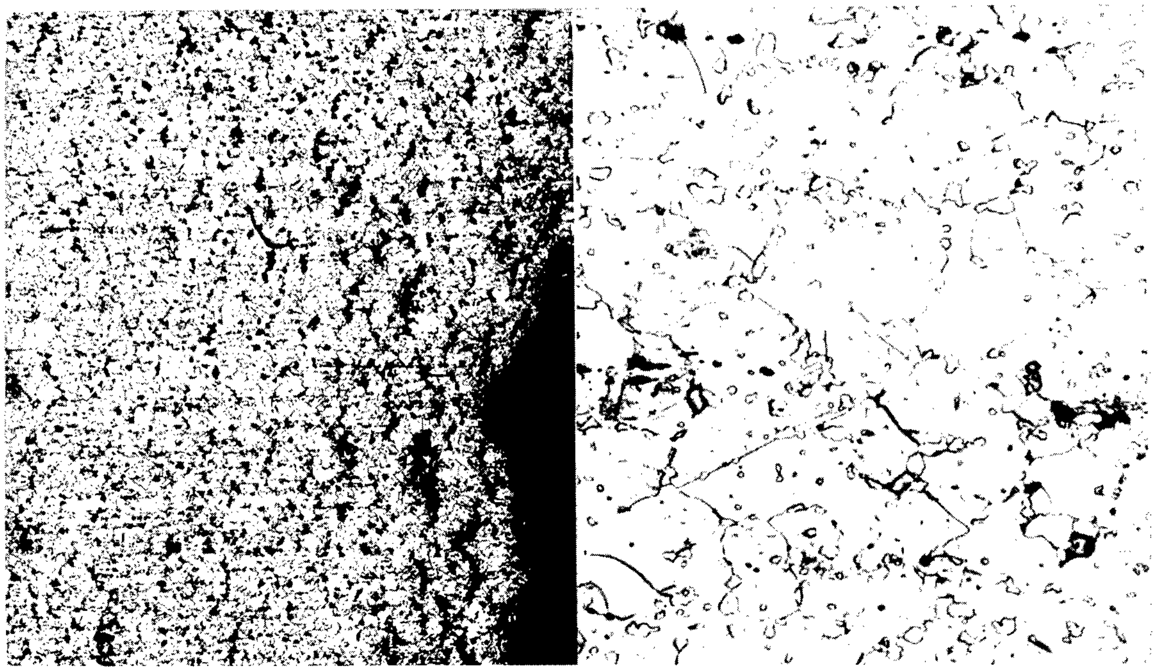
Figure 9.- Concluded.



100X 1000X

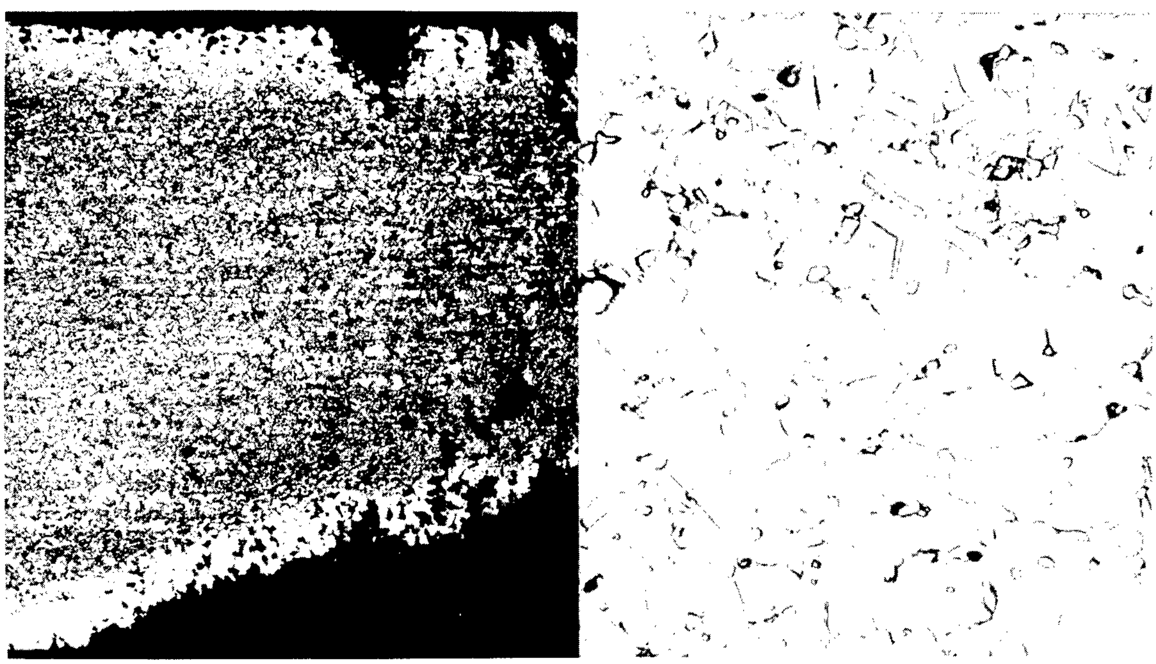
(a) Original microstructure

Figure 10.- Microstructures of sheet J-840 alloy. Acid ferric chloride etch.



Fracture - 100X Interior - 1000X

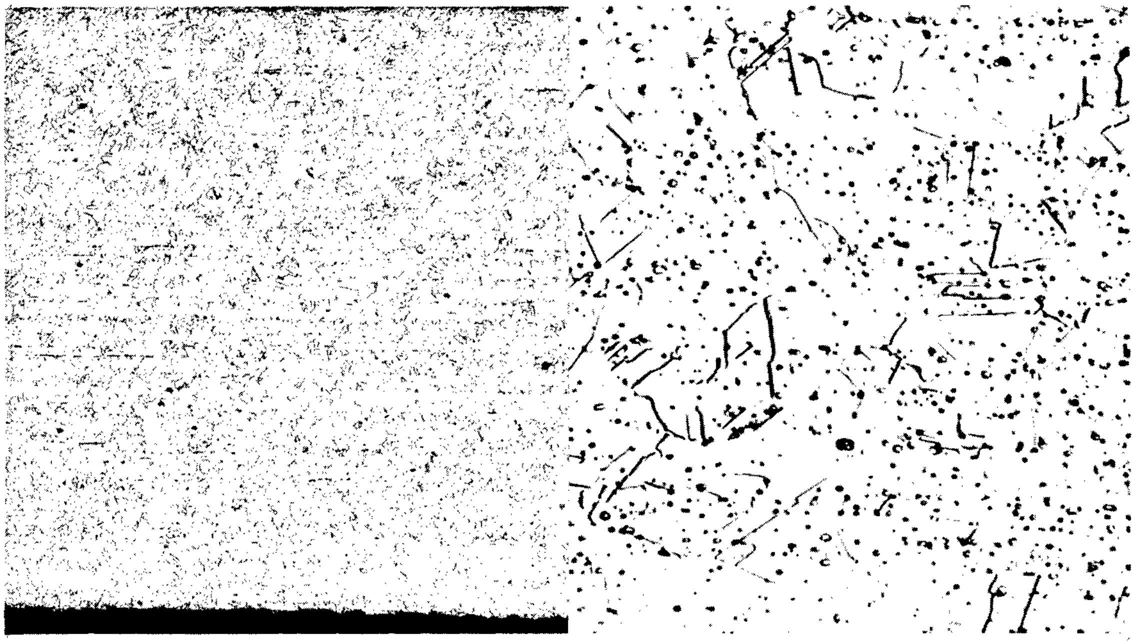
(b) 1133 hours for rupture at 1700° F under 3000 psi.



Fracture - 100X Interior - 1000X

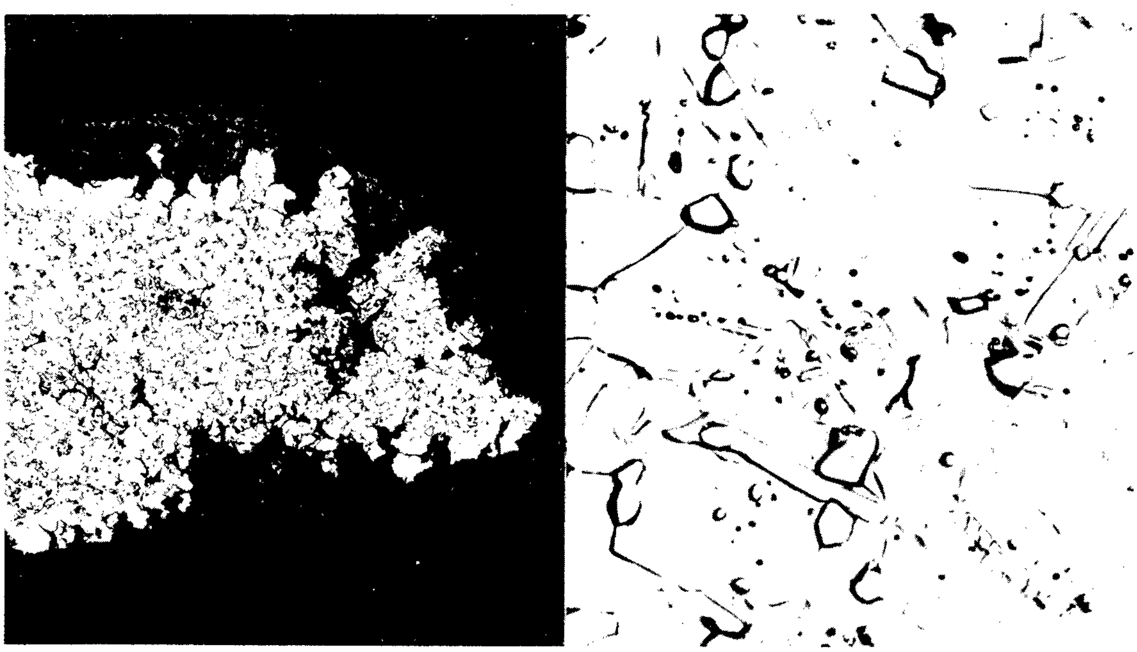
(c) 1309 hours for rupture at 1800° F under 1700 psi.

Figure 10.- Concluded.



100X 1000X

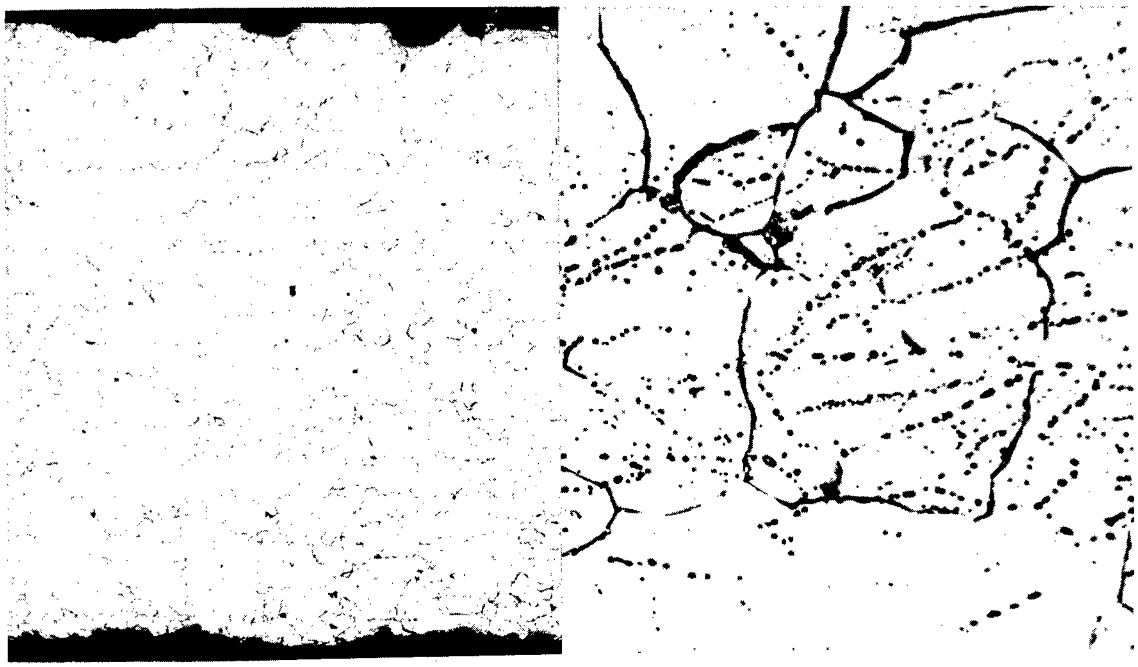
(a) Original microstructure.



Fracture - 100X Interior - 1000X

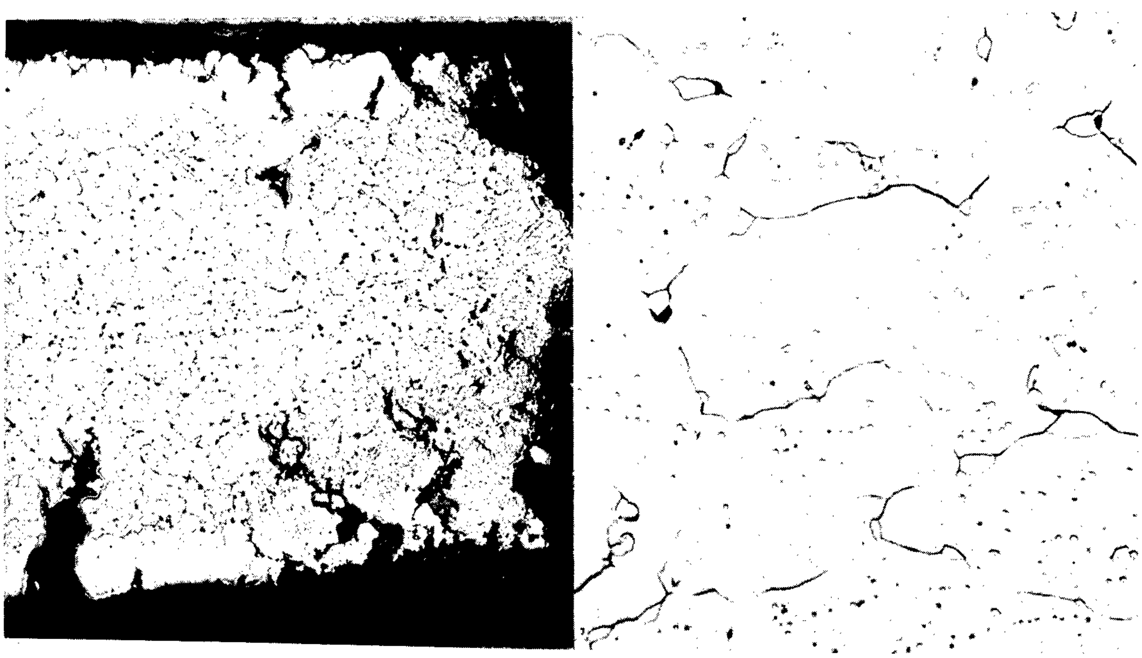
(b) 811 hours for rupture at 1700° F under 1500 psi.

Figure 11.- Microstructures of sheet type 330 alloy (15Cr-35Ni). Aqua regia in glycerine etch.



100X 1000X

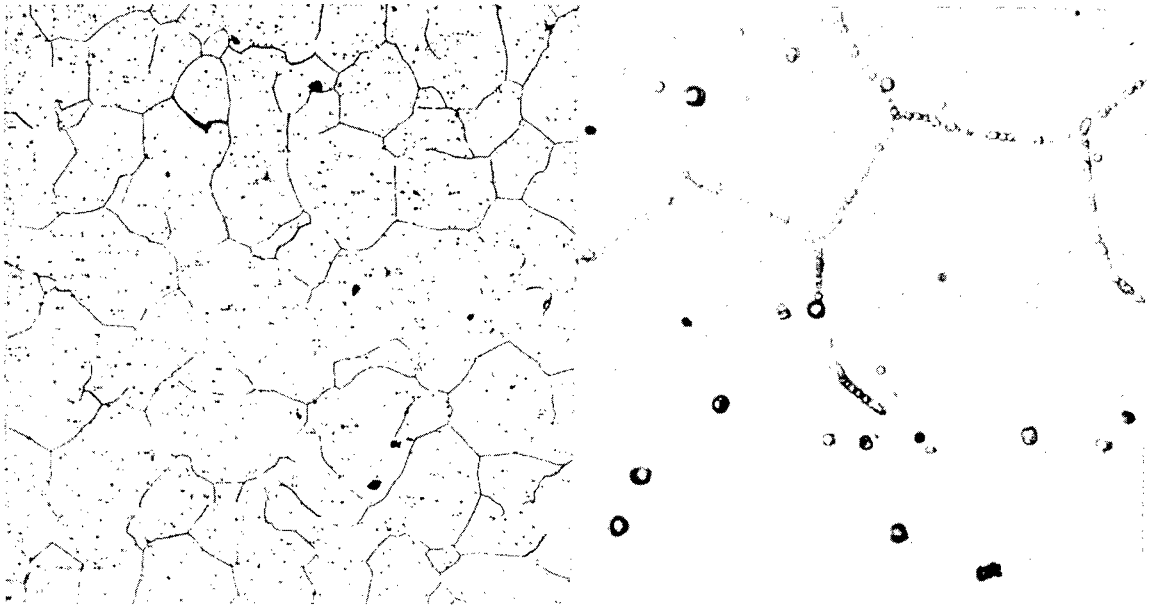
(a) Original microstructure.



Fracture - 100X Interior - 1000X

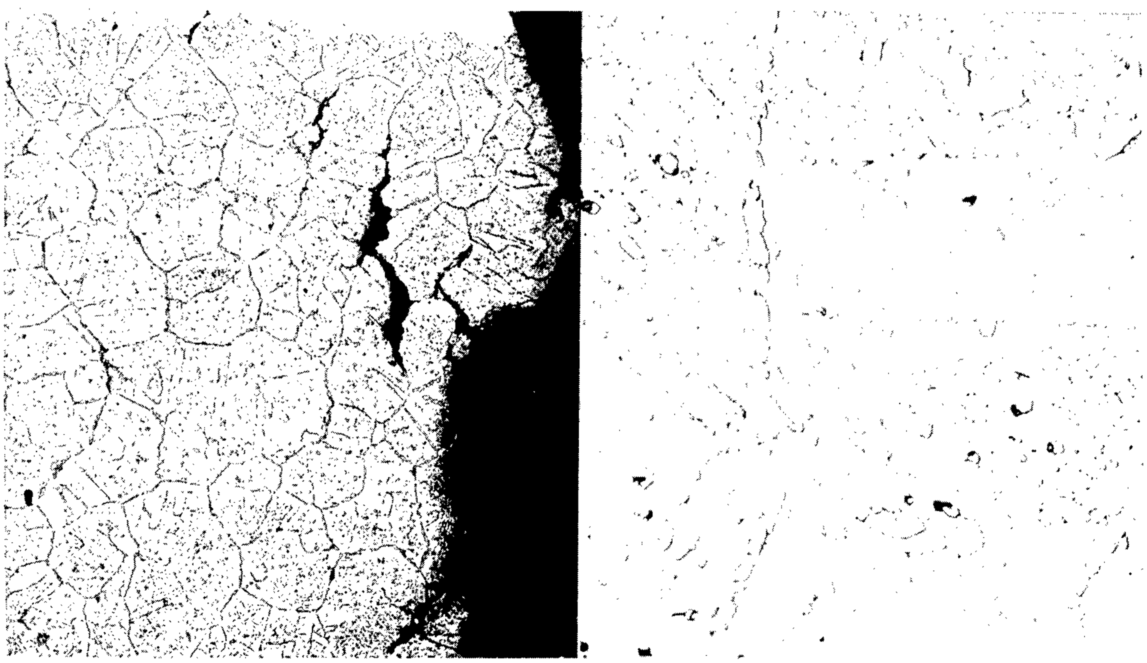
(b) 728 hours for rupture at 1700° F under 2500 psi.

Figure 12.- Microstructures of sheet type 310 alloy (25Cr-20Ni). Electrolytic chromic acid etch.



100X 1000X

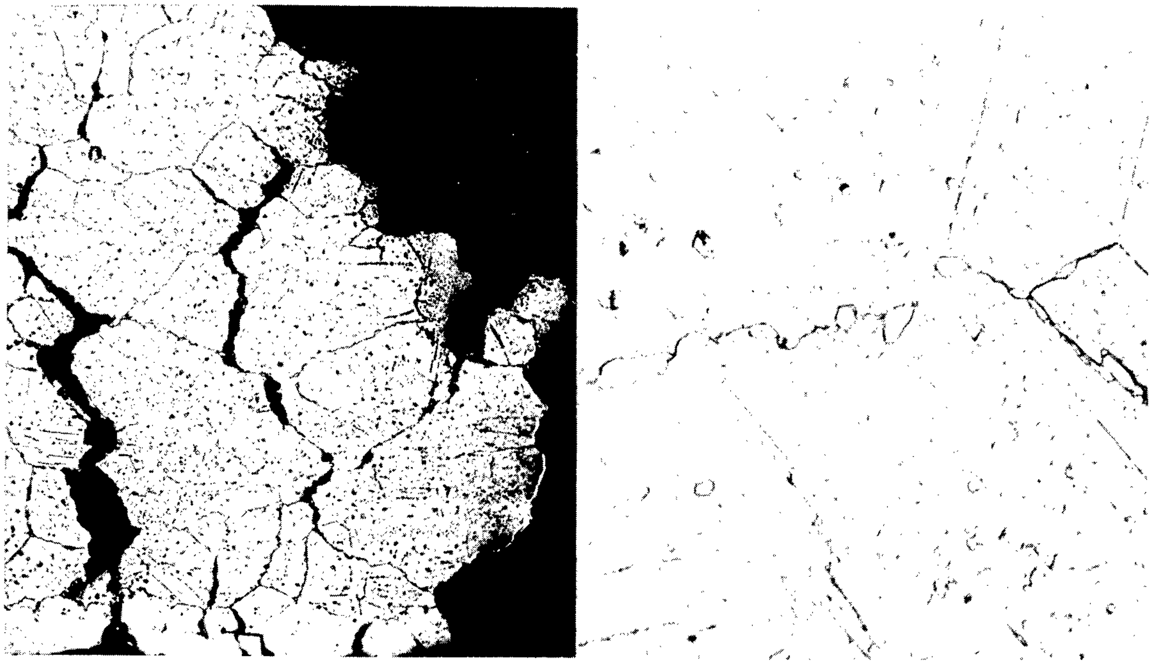
(a) Original microstructure.



Fracture - 100X Interior - 1000X

(b) 1124 hours for rupture at 1700° F under 3000 psi.

Figure 13.- Microstructures of sheet type 310S (AF-18) alloy (25Cr-20Ni+Si). Electrolytic chromic acid etch.



Fracture - 100X   Interior - 1000X

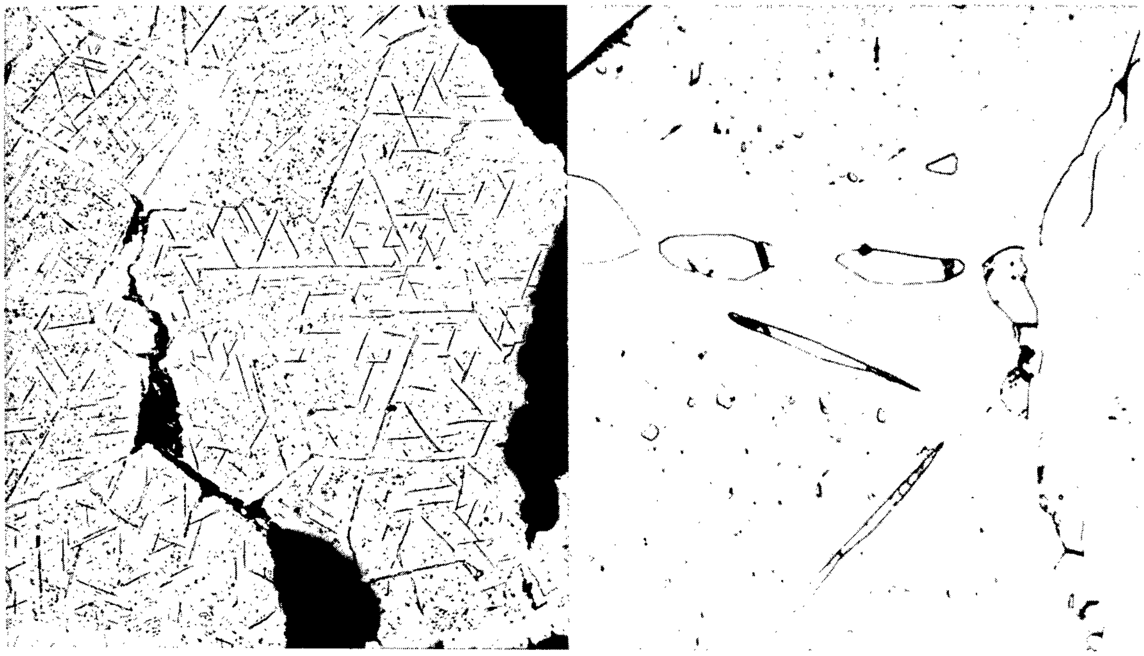
(c) 384 hours for rupture at 1800° F under 2200 psi.

Figure 13.- Concluded.

100X   1000X

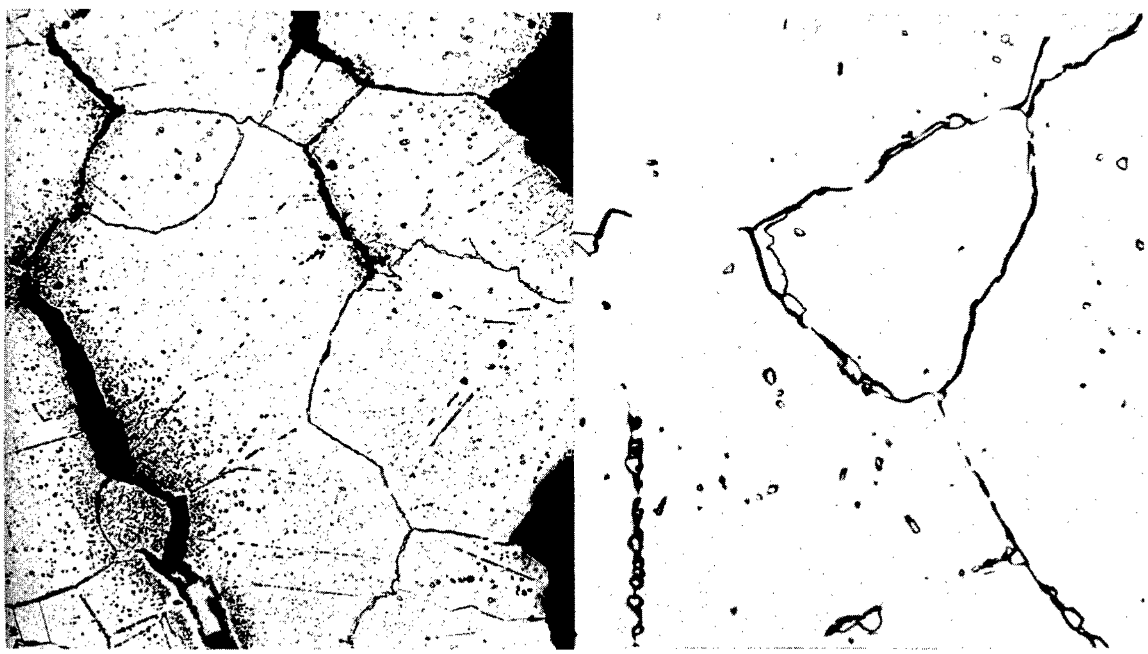
(a) Original microstructure.

Figure 14.- Microstructures of sheet type 310S (14626) alloy (25Cr-20Ni+Si). Aqua regia in glycerine etch.



Fracture - 100X Interior - 1000X

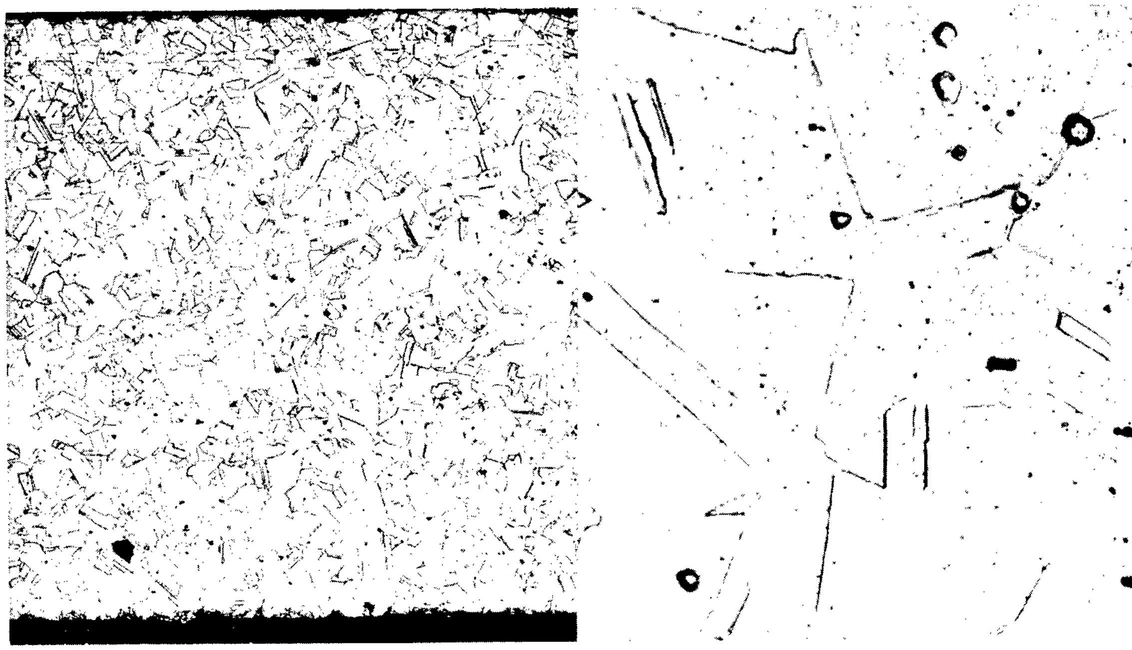
(b) 397 hours for rupture at 1700° F under 2400 psi.



Fracture - 100X Interior - 1000X

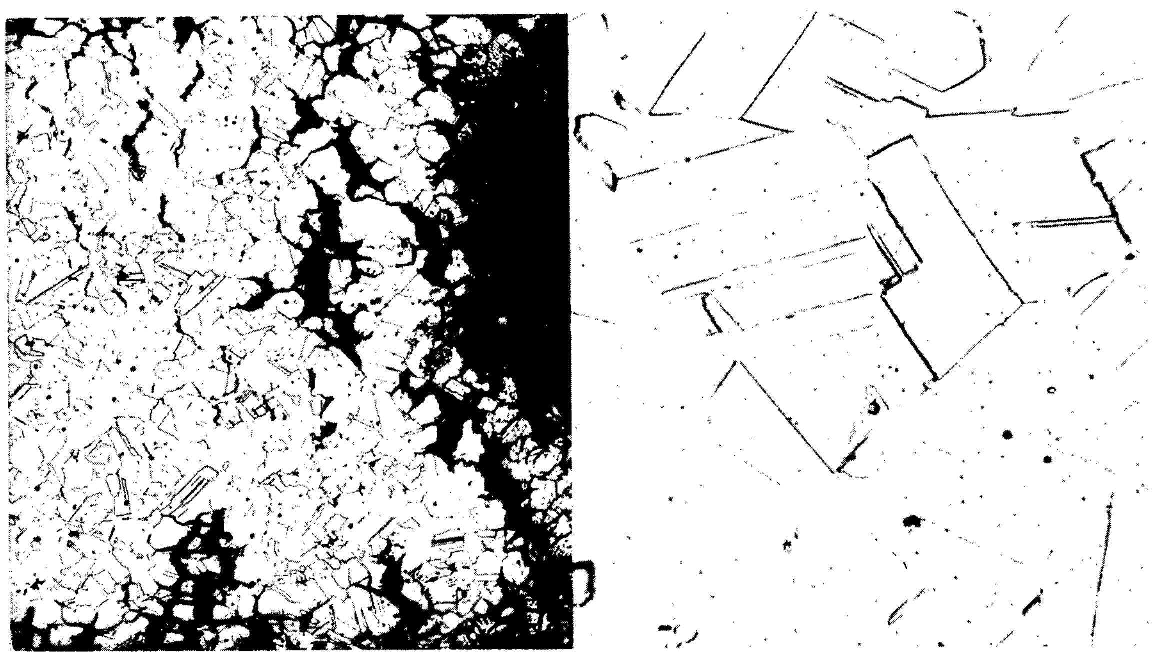
(c) 569 hours for rupture at 1800° F under 1700 psi.

Figure 14.- Concluded.



100X 1000X

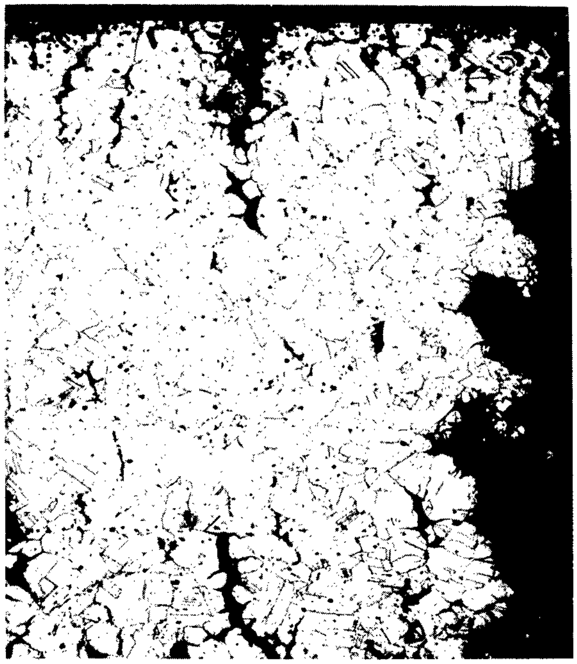
(a) Original microstructure.



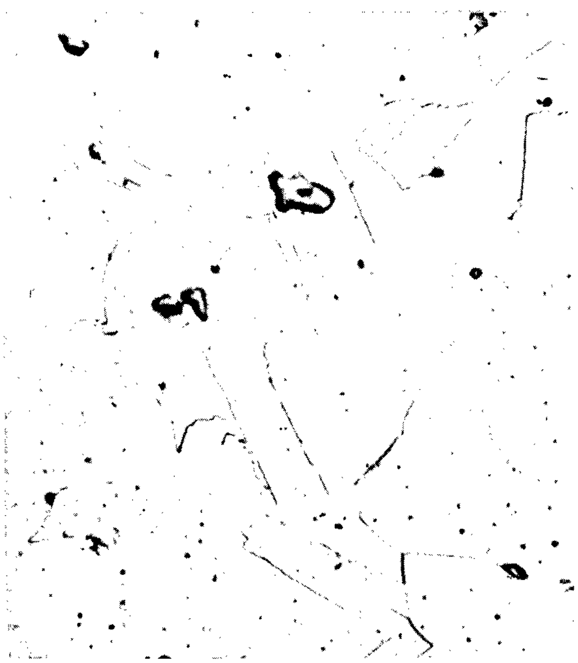
Fracture - 100X Interior - 1000X

(b) 277 hours for rupture at 1700° F under 2500 psi.

Figure 15.- Microstructures of sheet type 309 alloy (25Cr-12Ni). Electrolytic chromic acid etch.



Fracture - 100X



Interior 1000X

(c) 321 hours for rupture at 1800° F under 1800 psi.

Figure 15.- Concluded.